Coiling end of Tinplate Temper Rolling Mill

Richard Thomas

METAL WORKING AND HEAT-TREATMENT MANUAL

VOLUME I

By

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METAL WORKING

AND

HEAT-TREATMENT MANUAL

Volume II

ALLOY STEELS, CAST IRON AND
NON-FERROUS METALS

Volume III

SURFACE HARDENING PROCESSES

Volume IV

FURNACES AND PYROMETRY

Preface

THE need to urge engineers to take a more active interest in the metallurgical study of the metals and alloys which are available for their use, is becoming less acute.

Authorities responsible for providing courses of training for the engineers of to-day and to-morrow, appear to have recognized the importance of finding room in their curricula for the subject of Metallurgy.

This is not the place in which to discuss the adequacy or otherwise of the provision thus made; it is at least gratifying to realise that an all-important link is at last being forged between engineers and metallurgists, thus establishing a liaison which has every prospect of growing closer and stronger to the mutual benefit and advantage of the engineering and metallurgical sections of industry.

In no field is there greater need for this liaison than in that of the treatment which metals and alloys receive before being delivered to the engineer.

It had been the intention of the publisher and the author to produce a book dealing with the Mechanical and Heat Treatment of Metals. The vast scope of such a project and the impracticability of embracing both fields in a single volume, led to the decision to treat them separately, giving first consideration to heat-treatment.

The present volume, which is the first of four, deals almost exclusively with carbon-steels.

Although a knowledge of the principles of physical metallurgy will be an advantage, it is hoped that readers not possessing such knowledge will be able to profit from a study of the matter set forth in this volume.

Forthcoming volumes will deal successively with Alloy Steels, Cast Iron and Non-ferrous Alloys (Vol. II); Surface-hardening Processes (Vol. III); and Furnaces and Pyrometry (Vol. IV).

Prepared under the stress and strain of war conditions, with all the frustrating events inseparable therefrom, and under the additional handicap of illness, some limitation has been unavoidably set to the scope of this work and to the possibility of avoiding errors. The author would therefore be grateful for any constructive suggestions for the improvement of future editions.

He desires to acknowledge gratefully the assistance which he has received from : (a) The City of Birmingham Gas Department for permission to publish the description of one of the heat-treatment sections controlled by the Department's Industrial Research Laboratories; (b) Messrs. Chapman & Hall and Professor R. G. Batson for permission to make use of the text and illustrations in their text-book " Mechanical Testing," and various other authorities mentioned in the text; (c) Mr. S. G. Cope for valuable suggestions during the reading of the proofs.

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Heat-Treatment of Metals

VOLUME I

CARBON STEELS

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American Standard Specifications for Carbon and Alloy Steels. (Compositions only.) British Standard Specifications for Steel Bars and Forgings. D.T.D. and British Aircraft Standards. (Compositions and Mechanical Properties.)

Chapter 1

HEAT TREATMENT OF CARBON STEELS

INFLUENCE OF COMPOSITION AND TREATMENT ON THE STRUCTURE AND PROPERTIES OF STEEL

The two chief reasons why steels are the most important constructional materials are as follows :—

(1) Many of the initially good properties of the metal iron are enhanced by the addition of carbon, variations in the percentage thereof producing a very wide range of useful properties.

(2) The wide range of properties obtained by variations in composition is widened still further by applying different kinds of treatment, *e.g.* hot-working, cold-working and heat-treatment. Further extension is obtained by the addition of alloying elements such as nickel, chromium, etc.

Composition.

Theoretically the term steel is applicable to alloys of iron and carbon containing not more than 1.7% of the latter. In practice, however, such alloys are not likely to be accepted as steels unless they can be successfully hot-worked. Providing that this condition be fulfilled, the carbon in a steel for special purposes may exceed 1.7%. Malleability is largely affected by the presence of elements other than carbon. This property is most affected at high temperatures by the presence of sulphur, the adverse effects of which are corrected by the addition of manganese.

The chief influences of carbon on the properties of iron (the degree of influence varying with the percentage of carbon) are as follows :—

(1) Lowering of the freezing point, *e.g.* :—

Freezing point of pure iron	1535° C.
Freezing range of 1.7% C. steel ..	1400–1135° C.
Freezing point of 4.25% C. alloy ..	1135° C.

(2) Lowering of welding and forging temperatures.

(3) Increase of hardness and resistance to wear.

(4) Increase of elastic limit, yield point, and tensile strength. Decrease in percentage elongation and percentage contraction of area, *e.g.* :—

	Tensile strength, tons/in.^{2}	Elongation %
Pure iron	18·6	34
Normalized steel, 0·30% C.	37·4	24

(5) Lowering of malleability and ductility, both hot and cold.

(6) Lowering of toughness.

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(7) Lowering of specific gravity.

(8) Lowering of electrical and thermal conductivity, *e.g.* the thermal conductivity of steel containing 0·85% carbon at 100° C. is 0·160 cal/cm²/cm/sec/° C. as compared with 0·208 for electrolytic iron at the same temperature.

(9) Lowering of resistance to corrosion.

(10) Alteration of appearance of fracture especially after heat-treatment.

(11) Lowering of magnetic permeability.

(12) Raising of electrical resistivity, magnetic retentivity and coercive force.

(13) Improvement of machineability.

(14) Conferment of the property of hardening by suitable heat-treatment.

Influences (1) and (2) are shown in Fig. 27, p. 43.

Influences (3) in Figs 2 and 3.

Influences (4) in Fig. 2.

It is with influence (14) that this book is mainly concerned.

Mechanical Properties of Carbon-Steel.

In slowly-cooled carbon-steels there are two phases, viz. : ferrite and cementite. In hypo-eutectoid steels ($<0·89\%$ carbon), these two phases are intimately mixed in the form of pearlite, which is known as a constituent, the other constituent being free ferrite, *i.e.* the ferrite phase unmixed with cementite.

In hyper-eutectoid steels ($>0·89\%$ carbon), the two phases are intimately mixed in the form of pearlite as before, but the other constituent is free cementite, *i.e.* the cementite phase unmixed with ferrite. The properties of ferrite and cementite are widely different. Ferrite has a Brinell hardness number of about 67, whereas that of cementite is about 600.

Ferrite is soft and tough ; cementite is hard and brittle. An intimate mixture of the two (pearlite) must of necessity possess properties intermediate between these two extremes. The percentage of carbon in cementite (Fe_3C) is 6·69. The percentage of cementite in a steel is

$$\frac{\text{carbon per cent.} \times 100}{6·69}$$

e.g. 0·89% carbon steel $0·89 \times 100$ 13·3%.

By using the factor 14·95, the percentage of cementite in any steel may be obtained thus :—

Per cent. carbon $\times$ 14·95.

Since the carbon in a normalized steel exists as iron carbide (Fe_3C), its powerful influences are more readily understood when carbon content is expressed in terms of iron carbide. Again, it must be remembered that in such steels, the iron carbide enters into a partnership with ferrite, as already pointed out, to form the constituent pearlite, which occupies a much higher volume percentage of the steel than iron carbide alone. Thus 13·3% carbide is the equivalent of 100% pearlite. To express carbon percentages in terms of pearlite percentages, the former should be multiplied by 112·36.

Thus :   0·445% carbon = 0·445 $\times$ 112·36 = 50% pearlite.

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The tensile strength of pearlite depends upon its condition (finely or coarsely lamellar).

Taking the tensile strength of ferrite as approximately 20 tons/sq. in., that of pearlite might be as high as 60 tons./sq. in.
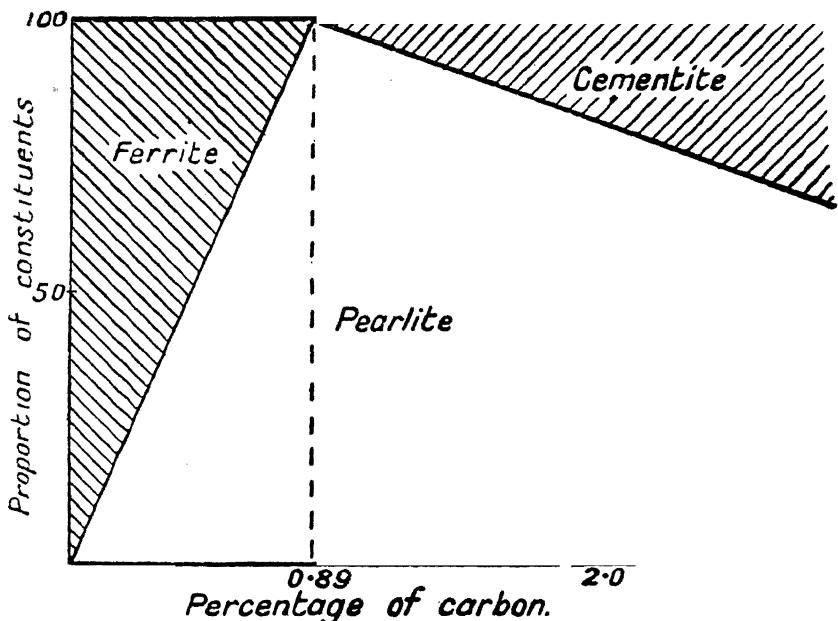

FIG. 1. RELATIVE PROPORTIONS OF CONSTITUENTS IN SLOWLY-COOLED (ANNEALED) CARBON STEELS. (*Gregory.*)

The tensile strength and hardness of hypo-eutectoid steels are proportional to the amount of pearlite, *e.g.* steel containing 0·445% carbon consists of 50% ferrite and 50% pearlite. Its tensile strength should be

$$\frac{20 + 60}{2} \text{ tons/sq. in.}$$

$$= \frac{80}{2} \text{ tons/sq. in.} = 40 \text{ tons./sq. in.}$$

The diagram (Fig. 1) indicates the relative proportions of ferrite and pearlite in hypo-eutectoid steels and of cementite and pearlite in hyper-eutectoid steels. These relative proportions vary with the composition of commercial steel. The carbon content of pearlite is lessened by manganese, nickel, etc. (*see* Alloy Steels, Vol. II). In order fairly to compare the mechanical properties of steels of various carbon contents, there should be no other variable but carbon content. Owing, however, to the inevitable differences in purity between low carbon and high carbon steels and in methods of manufacture, amount of mechanical treatment, etc., it is practically impossible to obtain a series of commercial carbon

steels in which all other conditions except carbon content are constant. Some indication of the trend of the influence of increasing carbon on mechanical properties may be obtained from the test-results plotted in Fig. 2.
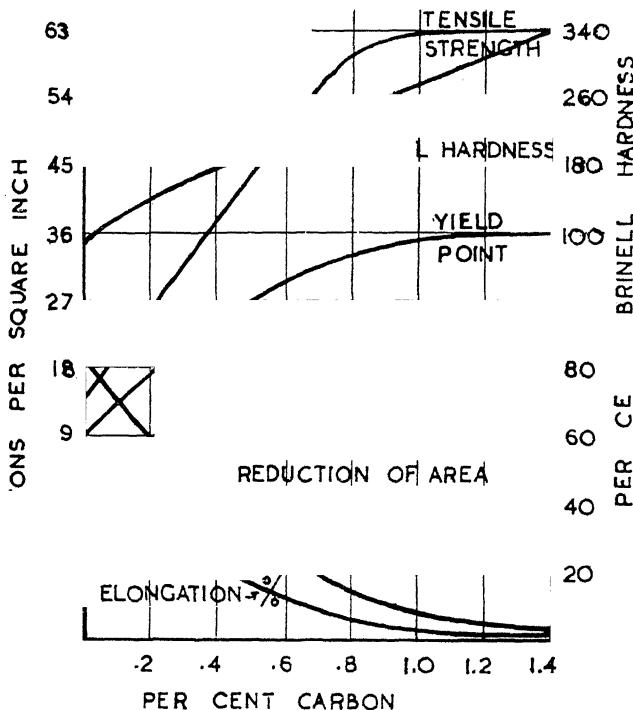
FIG. 2. VARIATION OF MECHANICAL PROPERTIES WITH CARBON CONTENT FOR NORMALIZED STEELS.

Heat-Treatment. Influence of Carbon.

Iron is a metal which undergoes phase-changes at certain temperatures. It is, like all other metals, crystalline, and, like the majority of metals, crystallizes in the cubic system. Its atoms are symmetrically arranged in a cubic pattern which is repeated over and over again. This is known as a space-lattice, each cubic space being formed by equi-distantly placed atoms, one of which is situated at each cube corner.

There is also an atom at the centre of each cubic space. In this close-packing of cubes (or cubic spaces) all of the same size, each cube corner must inevitably coincide with the corners of seven other cubes, and at every such point where eight cubic spaces meet, one atom only is located. Thus each corner-atom of each cubic space is shared by eight such spaces. Each cubic space has eight corners, but can claim only one-eighth of a share in the atom located at any one of these corners, *i.e.* it can claim " ownership " in

respect of eight eighths or one atom in all. Together with the one atom located at its centre, its full " ownership " is two atoms. (*See* Fig. 4.)

This arrangement is known as the " body-centred cubic space lattice " and is characteristic of iron below about 910° C. The critical change is at 937° C. on heating, the reverse change on cooling being at 910° C.

Above this critical temperature, the atomic arrangement is transformed, but not the crystalline system, which remains cubic. A new space-lattice is formed, in which there is still an atom at each corner of each cubic space,
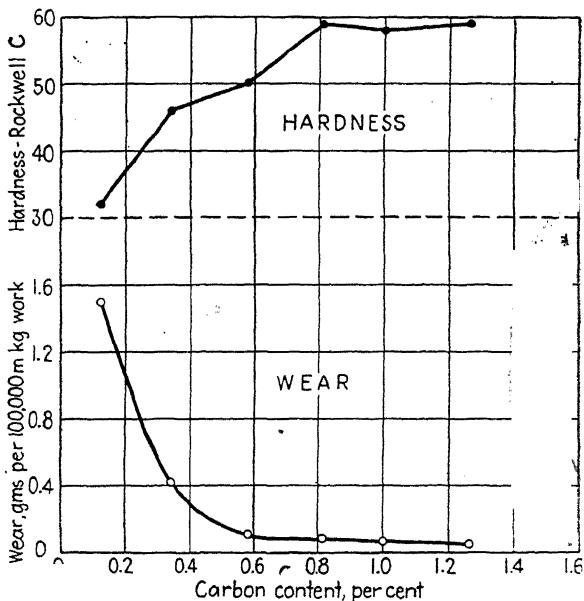

FIG. 3. INFLUENCE OF CARBON ON THE HARDNESS AND WEAR-
RESISTANCE OF HARDENED STEELS.
(*Rosenberg. U.S. Bureau of Standards. J. of Research, 1931.*)

but no atom at the centre thereof. Instead, there is an atom at the centre of each face of each cubic space or rather at the centre of the area which would be a cube face if the cubic space were a solid cube. Each cubic space now claims ownership to one atom representing eight $\frac{1}{8}$-shares in the corner atoms as before, and in addition, to a share in each of the six face-centred atoms. This share is one-half, since the face of an adjacent cubic space coincides with each face of the cubic space under consideration and can therefore claim joint ownership in the face-centred atom. Six half-shared are equivalent to three whole atoms, which, added to the one atom representing those at the corners, makes four in all (*see* Fig. 5).

If the cubic spaces of this so-called " face-centred cubic space-lattice " were the same size as those of the body-centred lattice which they have replaced, a very considerable shrinkage would have had to take place to enable the *same total number of atoms* to occupy half the original total space. But the cubic spaces are not the same size as before ; they are larger

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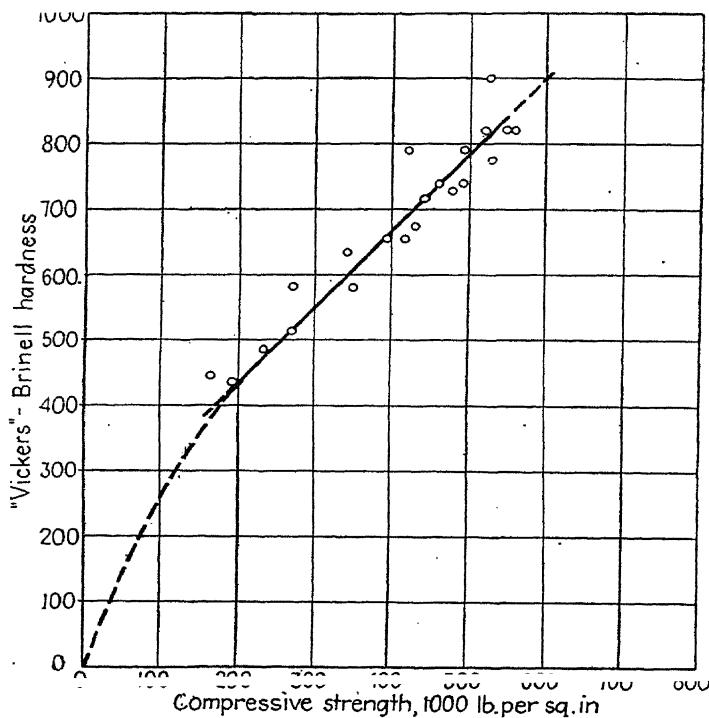
FIG. 3A.   RELATION BETWEEN HARDNESS AND COMPRESSIVE
STRENGTH OF HARDENED STEEL.
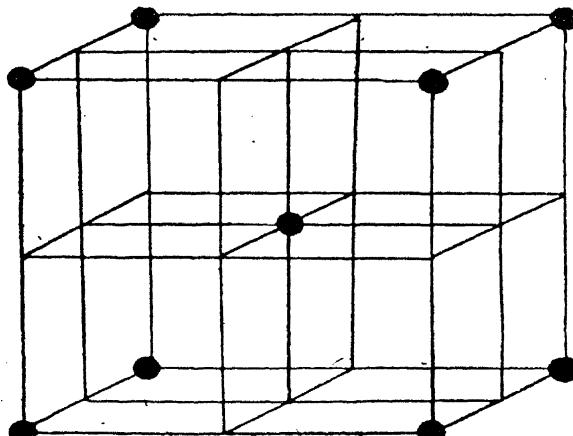
FIG. 4.   ARRANGEMENT OF ATOMS IN A UNIT CELL OF THE BODY-
CENTRED CUBIC SPACE LATTICE.

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but not quite twice as large, so that, in fact, a small diminution in volume (increase in density) does occur when the body-centred lattice (alpha phase) transforms to the face-centred lattice (gamma phase). There is also an *absorption* of heat accompanying this change. The reverse changes occur on cooling through the critical temperature, viz. :—a small increase in volume and an *evolution* of heat, resulting in a return to the atomic arrangement and density of the alpha-phase, which existed before heating. This cycle of changes occurs every time iron is heated and cooled through the critical temperature, at which there is always an arrest on heating, (Ac) and an arrest on cooling (Ar).

Iron, like other metals, consists of an aggregate of crystal-grains, of various sizes and shapes. When a specimen of pure iron is ground, polished and etched, the crystalline structure is revealed as an irregular network
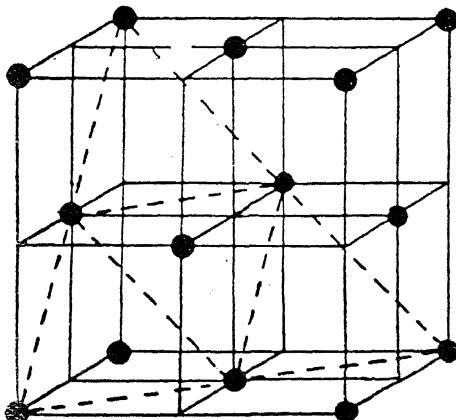
FIG. 5. ARRANGEMENT OF ATOMS IN A UNIT CELL OF THE FACE-CENTRED CUBIC SPACE LATTICE.

of dark lines, enclosing white areas. (Fig. 6.) These white areas are crystal grains of pure iron (ferrite) in section, the dark lines their boundaries. Though apparently structureless, each of these crystal grains is composed of atoms arranged in a body-centred cubic space lattice ; it differs from its neighbours only in the orientation of its space lattice. When these crystal grains undergo the transition from body-centred (alpha) to face-centred (gamma) internal atomic arrangement, their old structural arrangement of ferrite (α) disappears and in place of it a new structure appears, viz. :— that of crystal-grains of the gamma phase.

Whatever may have been the grain-size of the alpha phase, the recrystallization results in an aggregate of much smaller grains of the gamma phase. This grain refinement occurs every time the alpha to gamma change takes place. When carbon is added to iron (at any temperature from that of freezing point down to 910° C.) the carbon atoms occupy, in the face-centred space lattice of the solid gamma phase, interstitial places, *i.e.* places in between those occupied by iron atoms (*see* Fig. 7). The carbon is said to be

in solid solution. The number of places occupied by carbon atoms increases with the percentage of carbon present. When the steel cools to the transition temperature, Ar₃, the transformation to the body-centred lattice (alpha phase) is begun by iron atoms.
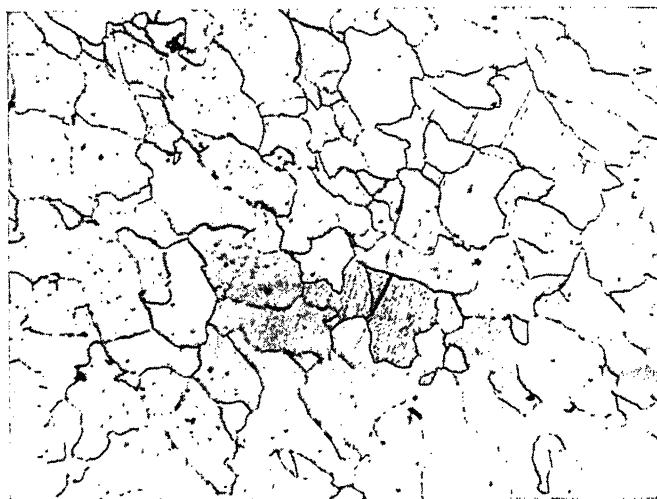
FIG. 6. STRUCTURE (FERRITE) OF COMMERCIALLY-PURE
IRON (ANNEALED). ×100.

As the separating alpha phase grows in amount, so the gamma phase diminishes and, for a time, the two different lattices co-exist, the body-centred alpha, free from carbon and the face-centred gamma in which all the carbon atoms have sought refuge, their concentration increasing until
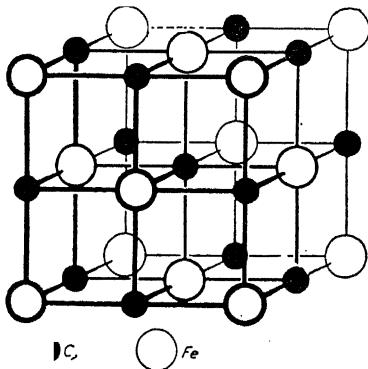

FIG. 7. POSSIBLE POSITION FOR
CARBON ATOMS IN A UNIT CELL
OF GAMMA IRON (AUSTENITE).

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it reaches saturation. The temperature has now fallen to about 720° C. and, at this temperature (Ar$_1$) the last structural change occurs.

What remains of the face-centred lattice disappears, most of the iron atoms separating into lamellæ of body-centred alpha iron, leaving each carbon atom a legacy of three iron atoms, this combination with its molecular formula Fe$_3$C., (iron carbide) and its own special space lattice, establishing itself as lamellæ alternating with the alpha iron lamellæ. This lamellar aggregate is an intimate mixture of alpha iron and iron carbide, the scientific name of which is eutectoid, the earlier and more popular name being pearlite, so called, because on its smooth polished etched surface it causes incident white light to be split up and to present a play of colours resembling mother-of-pearl. So in steels containing less than 0.89% carbon there are always two constituents, viz. :—alpha iron (ferrite) and an intimate mixture of alpha iron and iron carbide known as pearlite. (Fig. 8.) This applies to

FIG. 8. HYPO-EUTECTOID STEEL (0·35% C.) FERRITE (LIGHT) + PEARLITE (DARK).

steels which have not been subjected to any acceleration in the rate of cooling from temperatures above the Ar$_1$ point. A steel which contains $e.g.$ 0.89% carbon (this carbon content may be less than 0.89% if manganese or certain other elements are present) will consist of an aggregate of crystal-grains of austenite (solid solution of carbon in gamma-iron) above the Ar$_3$ critical-point. Carbon lowers the temperature of the Ar$_3$ point and, at this composition, Ar$_3$ and Ar$_1$ points coincide (at about 720° C.). At this temperature austenite is saturated with respect to both iron and carbon. Both are simultaneously precipitated and the place of each crystal-grain of austenite is taken by a crystal-grain of pearlite, in each of which alpha iron and iron

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carbide are inter-stratified in regularly alternating plates or lamellæ. Thus, in the place of an aggregate of crystal-grains of austenite, a new aggregate of crystal-grains of pearlite appears. (Fig. 9.) There is no separate ferrite constituent such as appears in the steels containing less than 0·89% carbon.

If the steel contains more than 0·89% carbon, a new separate constituent appears, viz. :—iron carbide, which separates from the austenite solid solution in the same way that ferrite separated in the steels containing less than 0·89% carbon. (Fig. 10.) Just as the temperature of separation of ferrite (Ar$_3$) increases with the amount of ferrite separating (i.e. with decreasing carbon content) so the temperature of separation of iron carbide increases with the amount of carbide separating (i.e. with increasing carbon content).

The residual austenite, after the separation of excess ferrite or iron carbide is complete, has always the same carbon concentration (viz. :—0·89% carbon) and its behaviour at the final critical point (Ar$_1$), is precisely the same
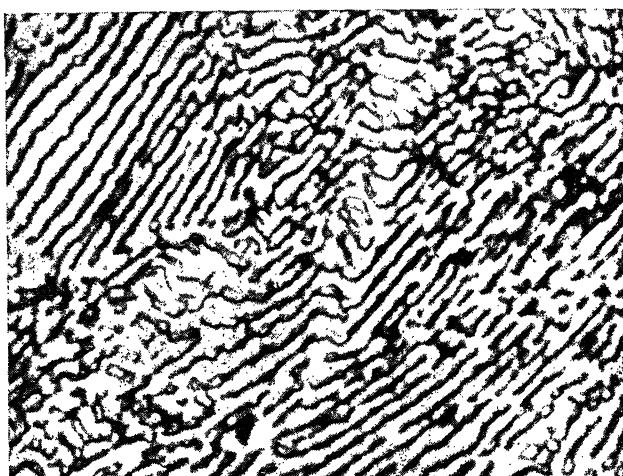
FIG. 9. EUTECTOID STEEL (0·89% C.) PEARLITE (LAMELLAR).
×2000.

as the behaviour of the 0·89% carbon steel which, at this temperature, changes to an all pearlite structure. Thus, whether the preliminary separating constituent is ferrite or iron carbide, its companion constituent will be pearlite.

Pearlite is harder than ferrite, so that hardness varies directly with carbon content in slowly-cooled steels. If the rate of cooling is very slow, the lamellar structure of pearlite may be readily recognized under the microscope at relatively small magnifications. The faster the rate of cooling the higher will be the magnification necessary in order to resolve the lamellar structure of the pearlite grains. This increasing degree of fineness of the ferrite and carbide lamellæ in the pearlite is responsible for increasing hardness and strength in the steel.

Steel is harder and stronger than iron by virtue of the presence of pearlite

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in any condition, but the degree of increase in hardness and strength is greater, the finer the internal structure of each pearlite grain.

A normalized steel consists of an aggregate of crystal grains, ferrite mixed with pearlite, pearlite alone or iron carbide mixed with pearlite. Hardness and strength, toughness and ductility, are affected by the number of these crystal-grains per unit volume, the best combination of properties being associated with finely-crystalline aggregates, *i.e.* structures in which there is a very large number of crystal-grains per unit volume. The greater this number the greater is the total area of inter-crystalline contacts.

The cohesion between crystal-grains is greater than the internal cohesion of any single one of them. A single crystal of iron, for instance, has a tensile strength of ten tons/in.^{2} whereas an aggregate of much smaller crystal-
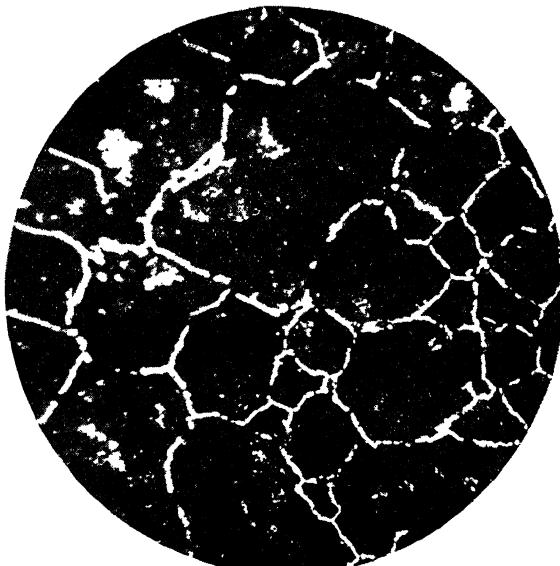
Fig. 10. Hyper-eutectoid Steel. Pearlite (Dark) + Cementite (Light). ×500. Slowly-cooled.

grains (75 per mm^2) has a tensile strength of eighteen tons/in^2. One of the objects of normalizing a steel is to convert a coarsely-crystalline structure into a finely-crystalline structure with its accompanying superiority in mechanical properties.

Thus, as the number of grains per unit volume increases, the resistance to stress increases. The overall superiority in mechanical properties of a medium carbon steel, compared with pure iron, is due to the uniform blending of a ductile constituent, ferrite, with a hard strong constituent, pearlite. In the contributions they make to the properties of the steel, each is complementary to the other.

Softness and plasticity will be greater, the greater the proportion of free ferrite. Hardness and strength will be greater the greater the proportion of pearlite.

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Since pearlite contributes so powerfully to the intensity of these las·
two properties, what would be the effect if the transformation from austenit‹
to pearlite were prevented. Would a medium carbon steel consisting o
equal proportions of ferrite and austenite be harder and stronger than th‹
same steel consisting of equal proportions of ferrite and pearlite ? Th‹
properties of such a steel are not known because it is not possible to produc
a steel so constituted. This is because the fastest rate of cooling which ca·
be devised, is not fast enough to prevent austenite from undergoing the initi‹
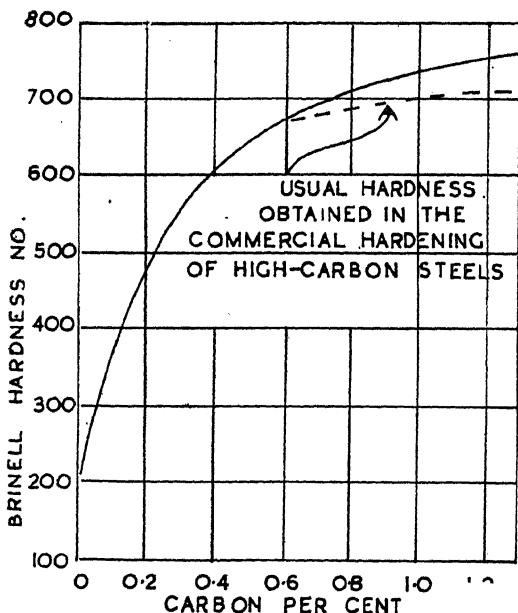
FIG. 11. VARIATION OF HARDNESS WITH CARBON CONTENT FOR
FULLY-HARDENED CARBON STEELS.

stages of the transformation to pearlite. If a medium-carbon steel which
has been allowed to cool slowly to a temperature between Ar_3 and Ar_1, a†
which it consists of equal proportions of ferrite and austenite, be suddenly
quenched in cold water, it will possess a structure quite different from the
ferrite + pearlite structure which it acquires when slowly-cooled. The
nature of the ferrite constituent is not affected by the rate of cooling ; that
of the other constituent is. It is neither austenite nor pearlite.

The dual precipitation of iron and iron carbide lamellæ from the austenite
solid solution, which occurs when pearlite is formed, is prevented by quench-
ing in water or other medium, which accelerates the rate of abstraction of
heat from the metal. But although pearlite is prevented from forming,
austenite is not retained unaltered.

The atomic arrangement (face-centred cubic space lattice) of austenite
undergoes an alteration and the perfect solubility of carbon atoms in gamma

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iron no longer persists. The result is a constituent which may be regarded as hybrid, its precise nature having not yet been fully established.

In Fig. 11 hardness is plotted against carbon content for steels in the hardened condition.

The upper curve represents the hardness of steels which have an entirely hardened structure.

The lower dotted curve represents the values usually obtained in commercial hardening practice.

No mention has previously been made of this new type of structure.

It is known as martensite and it possesses a characteristic microstructure. The characteristic appearance of martensite is seen in Fig. 12. It may be described as a network of triangular design. Its resemblance to a criss-cross arrangement of needles, placed parallel to the three sides of interlaced equilateral triangles has caused it to be described as acicular (from the Latin acicula—a small pin). Martensite is the hardest constituent obtainable in the iron-carbon alloys. Its hardness increases with carbon content, reaching over 700 Brinell at a carbon content *near* to that of a steel which, when slowly-cooled, consists exclusively of pearlite. When free iron carbide is also present, the hardness diminishes, because, although iron carbide is hard, it is less hard than the martensite with which it is associated.

The following figures are quoted for the Brinell hardness numbers of steels rendered martensitic by quenching:

TABLE I.

Carbon %	Quenching Temp. ° C.	Brinell Hardness No
0·35	870	550
0·50	840	675
0·70	800	713
0·90	750	683
1·18	800	683

These figures* agree closely with figures obtained under similar conditions by the author.

Concerning the real nature of martensite, finality has not yet been reached. It is certainly the result of an abortive attempt on the part of austenite to revert to the structure to which it would revert under conditions of slow-cooling, viz. :—the intimate mixture of ferrite and cementite (known as pearlite) produced by the dual precipitation of those phases at the Ar_1 critical point. This structure is the stable structure at atmospheric temperature.

The acceleration in rate of cooling prevents any change in the austenite until a very low temperature is reached. At this low temperature atomic movement is much less active. The normal regroupings into the two space-lattices of ferrite and cementite (in pearlite) cannot take place, but a certain amount of regrouping occurs, with the result that a freak constituent is formed, possessing the properties neither of austenite nor pearlite. Although martensite may be described as a freak, it is also by way of being a prodigy, for upon its formation and its intense hardness largely depend

* *They are probably affected by the presence of very small amounts of retained (or residual) austenite.*

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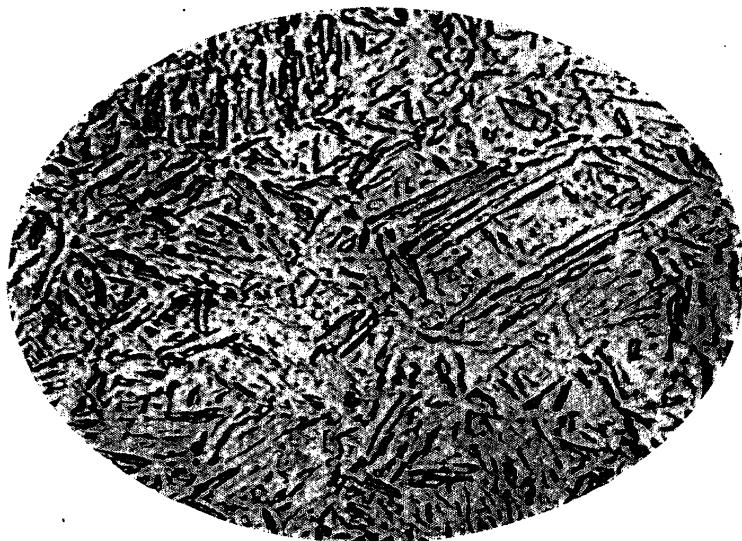
FIG. 12. STRUCTURE OF FULLY-HARDENED STEEL. MARTENSITE. (*Grossman.*) ₐ ×1000.

Range of positions for Fe atoms.

Probable positions for C atoms.
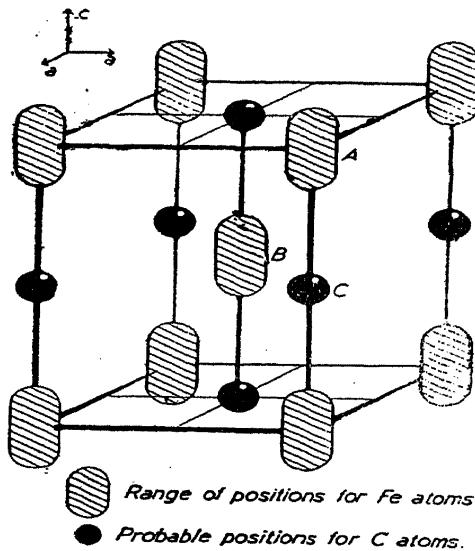
FIG. 13. PROPOSED LATTICE STRUCTURE OF MARTENSITE. (*Petch.*)

the enormous advances which have been made in industry. Without it, the material progress made by mankind could not have been achieved.

N. J. Petch* has suggested, from X-ray evidence, a possible explanation of the structure of martensite. By occupying interstitial spaces between the lattice positions taken up by iron atoms in the body-centred cubic space-lattice of alpha iron, the latter are displaced. The cubic cell becomes
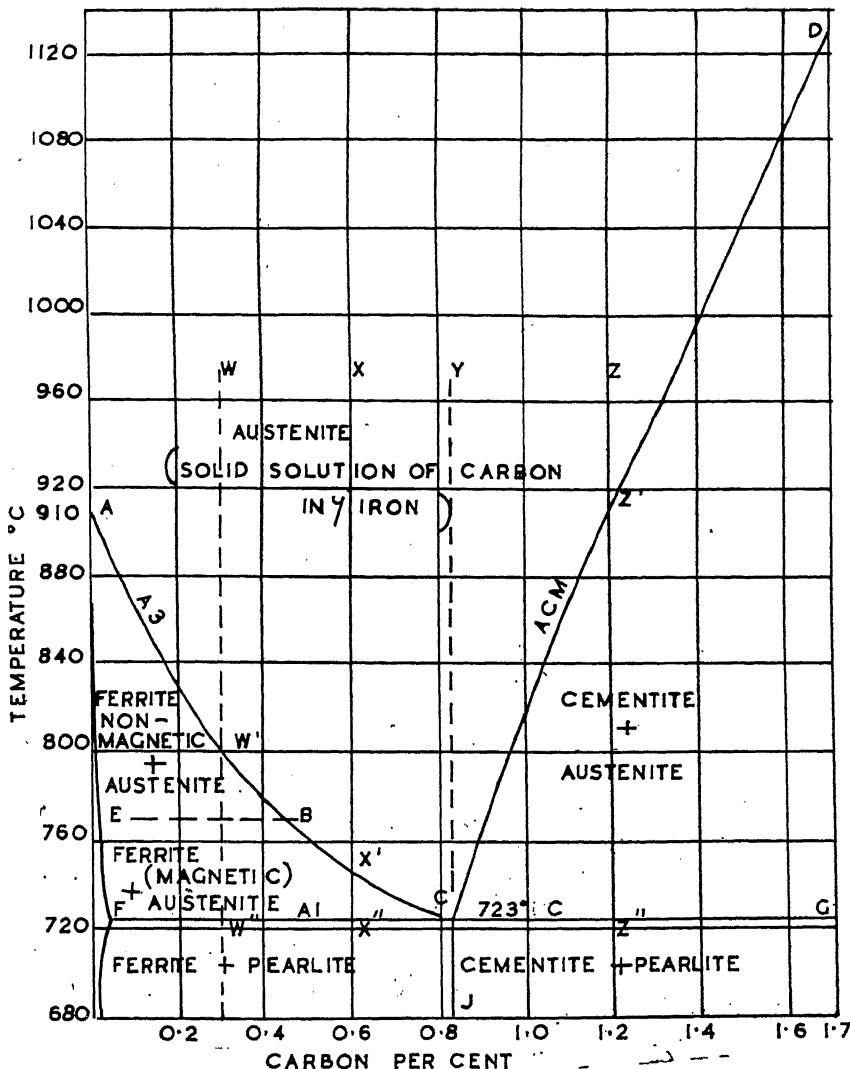
FIG. 14. PART OF THE IRON-CARBON CONSTITUTIONAL DIAGRAM.

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tetragonal, *i.e.* one of the three equal axes (the C axis) of the cubic cell becomes longer (by as much as 6%). The actual location of any carbon atom is midway between the iron atoms on the long edge of the martensite tetragonal cell or at the centre of a face perpendicular thereto. If all these interstitial positions were occupied by carbon atoms, the result would be as shown in Fig. 13.

There is, however, never enough carbon in solution to fill all these positions, which would require 2·0 carbon atoms per unit cell. The maximum carbon content possible in austenite can only provide 0·16 carbon atom per unit cell, which works out at one position in twelve available.
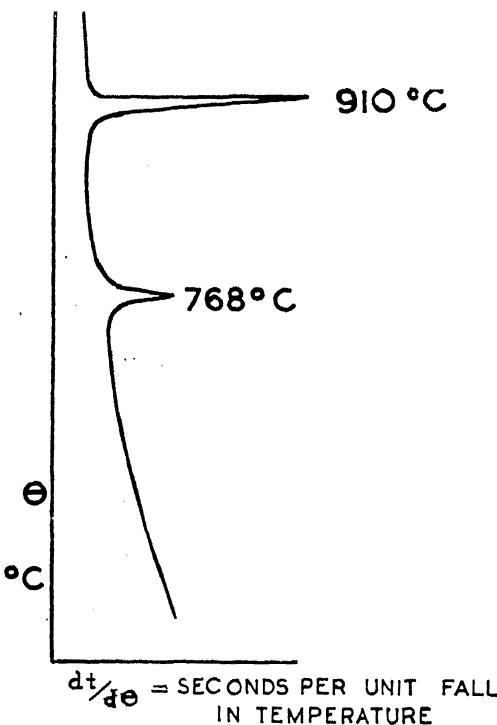
FIG. 15. COOLING CURVE OF PURE IRON.

These positions are crystallographically equivalent, each one being at the centre of an approximately octahedral cell, formed by six iron atoms.

In Fig 13 the black balls C represent possible carbon positions, and the elongated regions, A, B, the range of positions occupied by iron atoms.

If this explanation be correct, the hardness and brittleness of martensite would be associated with lattice-distortion and the triangular pattern of martensitic structures with the approximately octahedral disposition of the iron atoms, interlocked with a "shadow" octahedral arrangement of carbon atoms.

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The Iron-Carbon Constitutional Diagram.

In Fig. 14, the constitution of carbon-steels is shown by plotting temperature (in degrees Centigrade) on the vertical ordinate and carbon percentages on the abscissa. Two important arrest points on the vertical ordinate representing pure iron are obtained from time-temperature cooling curves (Fig. 15).

Ar_3 910° C. gamma $\longrightarrow$ alpha iron

Ar_2 768° C. acquisition of magnetic properties on cooling.

In addition to these, there is, at 695° C., when carbon is present, a third point (Ar_1) which represents the transformation of residual austenite to pearlite (*see* Fig. 16). Taking the first point, Ar_3, it will be noticed that the effect of adding carbon is a lowering of the temperature as shown by the sloping line AB. The second point is not lowered by the addition of carbon until, as shown by the horizontal line EB, it is joined by the falling Ar_3 point at 0·5% carbon. Henceforward, those two points Ar_3 and Ar_2 are inseparable and any point on the sloping line BC is represented by the symbol $Ar_{3—2}$. The change in magnetic properties at Ar_2 does not affect the constitution, either as regards space-lattice or microstructure.

The third point (Ar_1) occurs at constant temperature, as shown by the horizontal line FCG. All the lines representing the relations between critical point temperatures and carbon content are situated slightly higher when they are determined by noting the arrest points on time-temperature *heating*-curves. Points thus determined are called Ac points.

The diagram may now be utilized to study separately the behaviour of different classes of carbon-steel, *e.g.* low, medium and high carbon. A steel representing each of these types will be examined with regard to its behaviour (*a*) when heated, (*b*) when cooled.

(*a*) *Heating.*

0·30% *Carbon.*—When this steel reaches the temperature of the eutectoid horizontal FCG at the point W″, it consists, as it did at atmospheric temperature, of a little more than one-third pearlite and rather less than two-thirds ferrite. Immediately the critical point Ac_1 (temperature of horizontal FCG) has been passed, the pearlite has changed to austenite by virtue of the ferrite and iron carbide (cementite) lamellæ having dissolved each other. Henceforward as the temperature rises, the austenite (thus formed) grows in amount by feeding on the contiguous ferrite crystal-grains.

At the temperature of the line EB (Ac_2) what remains of the ferrite becomes non-magnetic. As the temperature rises still higher, this non-magnetic ferrite is gradually dissolved by the austenite, the carbon content of which has been steadily diminishing from 0·89% at Ac_1 until, at W′, the last traces of free ferrite are absorbed and the carbon content of the austenite now coincides with that of the steel itself, viz. :—0·30%. This final transition temperature (Ac_3) marks the completion of the change :— alpha space lattice of ferrite to the gamma-space lattice of austenite, which is the stable phase above Ac_3. If this steel had been quenched at a temperature between W″ and W′ (Ac_1 and Ac_3) it would have been found to consist of ferrite and martensite. If quenched above W′ it would consist entirely of martensite. The former condition would be harder than the normal

ferrite +pearlite condition which it possessed before heating ; the latter condition (exclusively martensite) would be still harder.

0·6% *Carbon.*—This steel consists of approximately two-thirds pearlite and one-third free ferrite (if in the normalized condition) at atmospheric temperature. At Ac_1 (line FCG) pearlite transforms to austenite, the carbon content of which will alter from 0·89% at $\times''$ to 0·6% when point x′ is reached, the dilution (as in the case of the 0·3% carbon steel) being due to the gradual absorption of the free ferrite. At x′ the loss of magnetism (Ac_2) as well as the alpha to gamma change (Ac_3) occur at the same temperature. One critical point ($Ac_{3.2}$) serves for both.

Quenching between $\times''$ and x′ will give a ferrite +martensite structure, while quenching above x′ will give an all-martensite structure (very much harder).

0·89% *Carbon.*—This eutectoid steel (all-pearlite) undergoes all three changes, viz. :—mutual solution of ferrite and iron carbide, loss of magnetism and the alpha to gamma change all at one temperature ($Ac_{3.2.1}$). Quenching above this temperature gives an all-martensite structure.

1·2% *Carbon.*—This steel consists, in the normalized condition at atmospheric temperature, of pearlite and free iron carbide (cementite). The behaviour of the pearlite will be the same as in the two steels already examined, viz. :—transformation to austenite at the temperature of the line FCG. The free cementite will be gradually dissolved by the austenite until at Z′ it finally disappears and above the line CD the steel consists entirely of austenite, of a carbon content (1·2%) richer even than the austenite of the eutectoid steel (0·89% carbon). Quenching between Z″ and Z′ results in a structure of cementite + martensite, whilst quenching above Z′ results in an all-martensite structure.

(b) Cooling.

0·30% *Carbon Steel.*—When slowly cooled from a temperature above line AB (say W), the austenite crystal-grains remain unaltered until temperature W′ is reached.

As previously indicated, the lines ABC, CD and FCG will be lower when plotted from cooling-curves and the critical points Ar_3 and Ar_1 will, of course, be situated on the lower lines. At W′ (Ar_3) ferrite separates from austenite, and continues to separate until W″ is reached, the residual austenite having become correspondingly richer in carbon, until a concentration of 0·89% carbon is reached.

The separated ferrite acquires magnetic properties at the line EB. At W″ the austenite transforms to pearlite and thenceforward no further changes occur. The ratio of pearlite to ferrite (1 : 2) is the same as before heating.

0·6% *Carbon Steel.*—Cooling from temperature X, the austenite crystal-grains undergo no change until at X′ ferrite separates. As this temperature is that of Ar_2 as well as Ar_3, the ferrite has magnetic properties at the temperature of its separative formation. The residual austenite becomes richer in carbon, changing from 0·6% to 0·89% between X′ and X″.

Its transformation to pearlite occurs at X″ (Ar_1).

Thenceforward no further change occurs. The ratio of pearlite to ferrite (2 : 1) is the same as before heating.

0·89% *Carbon Steel.*—Cooling from temperature Y, the austenite crystal-grains undergo no change until at C, they transform to pearlite crystal-grains, no ferrite having separated to cause any change in carbon-concentration. At C, the mutual solubility of carbon and gamma iron ceases, the latter changing to alpha iron, magnetic properties being simultaneously acquired. Thus the three changes, viz. :—gamma → alpha, (Ar_3), magnetic change (Ar_2) and cessation of solubility (Ar_1) all occur at one transition temperature (Ar_{3-2-1}).

The structure is now all-pearlite.

1·2% *Carbon Steel.*—Cooling from temperature Z, no change occurs in the austenite, until Z' is reached when iron carbide (cementite) separates
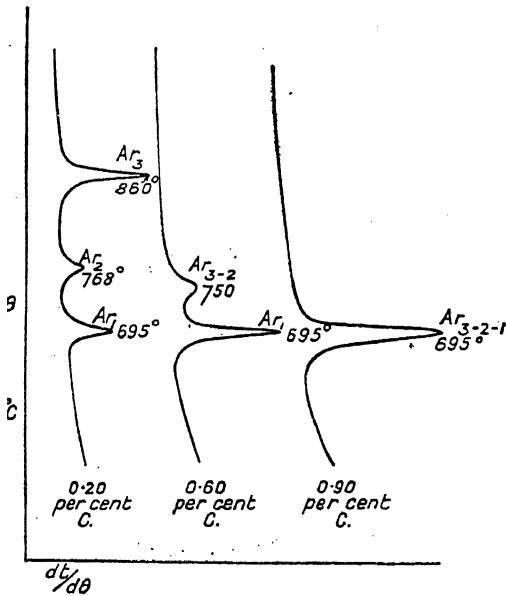
FIG. 16.   INVERSE-RATE COOLING CURVES FOR PLAIN CARBON
STEELS.   (*Gregory.*)

and, as the temperature falls, continues to separate, until the residual austenite, as the result of this loss, alters in carbon concentration from 1·2% to 0·89%.

Temperature Z'' has now been reached and the austenite transforms, as in all carbon steels at this temperature (Ar_1), to pearlite.   The structure is now cementite (iron carbide) +pearlite.

Typical cooling-curves are shown in Fig. 16.   There is, therefore, a definite relationship between the composition of the steel and the temperature at which it can be hardened.   The higher the carbon content the lower the temperature of hardening (*i.e.* full hardening).   To secure the maximum hardening effect, the steel, whatever carbon content up to that of the eutectoid (theoretically 0·89% carbon) must be at a temperature well within the austenite field, when quenched.

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With hypo-eutectoid steels (less than 0·89% carbon) not only must the free ferrite disappear before hardening, but a still higher temperature must be reached than that of Ac₃, which marks this disappearance.

The dotted line (in Fig. 27) represents the temperature at which steels should be quenched to ensure *full* hardening. When a steel is fully hardened its structure is wholly martensitic.

If the conditions obtaining during the hardening operation are such that the steel is not fully hardened, the microstructure will not be wholly martensitic. A new dark-etching constituent will make its appearance. This is known as troostite. If interwoven in the martensitic pattern it is called acicular troostite or the microstructure may be described as troosto-martensite (Fig. 17). It may also appear as a dark-etching intercrystalline network or as separate dark-etching nodular masses (nodular troostite) (Fig. 18). Troostite is softer than martensite but harder than pearlite. When examined under very high magnification it is seen to possess a lamellar structure.

Although more akin to pearlite than to martensite in properties and micro-structural appearance, troostite differs from pearlite in having a variable carbon-content, whereas pearlite has a constant carbon-content. It may be looked upon as incipient pearlite, containing less or more than 0·89% carbon.

The greater the proportion of troostite the less complete has the hardening operation been and the lower the hardness number of the steel. The different forms of troostite are shown in the microstructures of Figs. 17 and 18.

A steel which contains troostite after it has been hardened, is softer and tougher than the same steel which has been fully hardened to martensite. The latter condition, however, is preferable if tempering (especially at high

FIG. 17. ACICULAR TROOSTITE (DARK) IN MARTENSITE (LIGHT). ×500.

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temperatures) has to be done. Thus a well-tempered steel has a greater toughness if previously hardened to martensite (*i.e.* fully hardened) than if previously hardened to martensite and troostite.

When austenite first forms from aggregates of cementite and ferrite, it is not-homogeneous.* This inhomogeneity, attributable to both un-dissolved carbide and carbon concentration gradients, has a profound effect on the characteristics of the austenite → pearlite reaction. The rate of formation of pearlite is increased by the existence of inhomogeneity and in consequence, when the latter is sufficiently pronounced, the hardenability of the steel is lowered.

In addition to this kinetic effect, inhomogeneity can cause direct formation of spheroidite from austenite at temperatures just below A_{c1}.
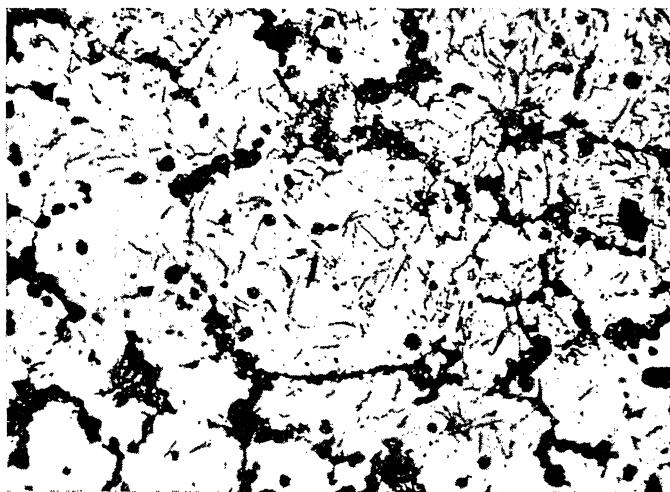

FIG. 18. MARTENSITE (LIGHT) + TROOSTITE (DARK). ×200.

Tempering.

The intense hardness of martensite, obtained when steel is fully hardened, is inseparable from brittleness. The operation of tempering has for its object the mitigation of this brittleness as well as the release of internal stresses. The equalization of internal stresses lessens the risk of cracking and this is the effect aimed at by tempering at low temperatures. Up to a certain temperature there is no sacrifice of hardness. There may be even a gain. The brittleness, as determined by bending or impact tests, gradually diminishes, the ductility increases, whilst the hardness, elastic limit and tensile strength fall off as the tempering temperature increases. (Fig. 20.)

Concomitant with these changes, there is a corresponding alteration in the microstructure, and indeed it is because of the internal structural change that the changes in mechanical and physical properties occur. As the result of an increase in temperature the atoms acquire greater freedom

* *Amer. Inst. Min. & Met. Eng. Tech. Publication. 1568, 16 pp. " Metals Technology," June, 1943.*

of movement, whereby they are enabled to release themselves from positions occupied under constraint, in favour of positions which are more in keeping with a condition of constitutional stability. The more completely the carbon atoms associate themselves with iron atoms in the form of iron carbide (cementite) the more closely is this condition of constitutional stability approached. The higher the tempering temperature the more rapidly is the precipitation of cementite achieved.
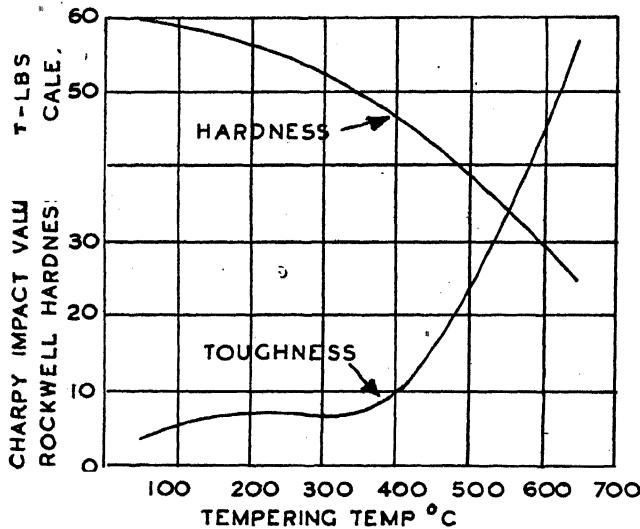

FIG. 19. EFFECT OF TEMPERING FULLY-HARDENED STEEL (CARBON 0·48%)

At the lower temperatures any structural change is difficult to recognize under the microscope, but as the temperature is increased the martensite begins to lose its acicular appearance. The microstructure becomes more and more stippled or granular as the result of the formation of numerous microscopically small particles of precipitated carbide. The name given to this constituent, or pseudo constituent, is sorbite (Fig. 21).

Naturally, as the highly complex arrangement of the atoms in martensite gives place to a structure in which particles of cementite are distributed in a highly-dispersed condition in what is practically a cushion of ferrite, a gain in ductility and toughness is only to be expected. The process of precipitation of fine particles of carbide is succeeded at still higher temperatures, especially after increased duration of time at those temperatures, by agglomeration.

The minute particles grow in size, their distance apart increases and the structural stiffening diminishes. This is described as "spheroidizing." Hence there is a critical degree of dispersion of these hard particles at which maximum resistance to impact is attained and beyond which that resistance falls off. In practice and for commercial tempering times, this stage is
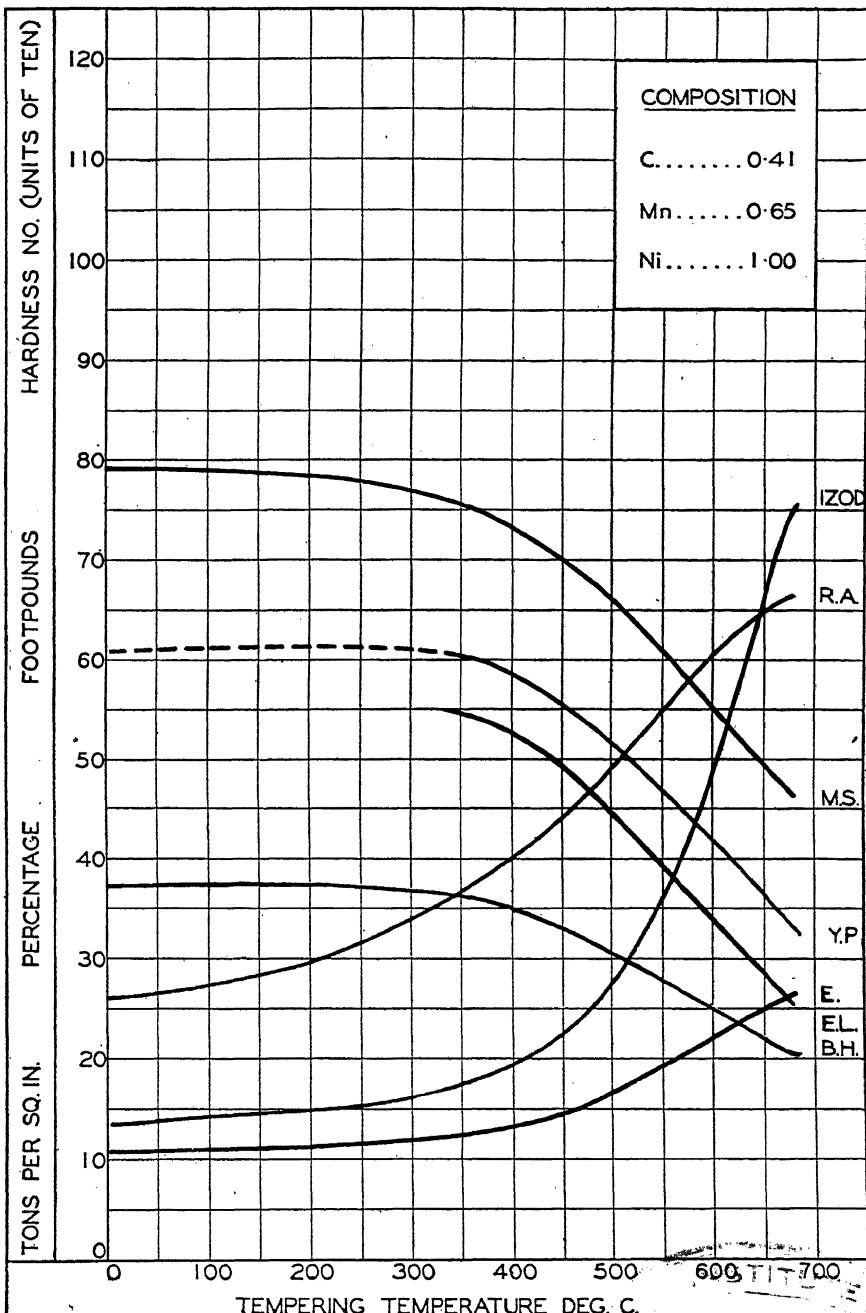
FIG. 20. INFLUENCE OF TEMPERING TEMPERATURE ON MECHANICAL
PROPERTIES OF HARDENED STEEL (WATER-QUENCHED)

reached at a temperature just short of the lower critical point (Ac_1). In order to obtain any specified combination of properties, the correct tempering conditions (time and temperature) must be fulfilled.

The old-fashioned method of judging the temperature by observing the colour of the oxide film formed on a pre-brightened surface of the steel is one which calls for skill and experience. The range of colours thus obtained with their corresponding temperatures together with the applications of the steel appropriate thereto are given in Table VIIIA, Chapter II. Control of tempering temperature, however, may be more precisely obtained by using heated oil or molten salt baths, the temperature of which is ascertainable by means of a thermometer or thermo-couple pyrometer.

This stricter control is justified by the greater regularity of the results, by the protection of the steel from oxidation and by the greater uniformity of heating obtainable in a heated liquid. The main factors governing the tempering operations are temperature and time, but although it is possible to obtain, by prolonging the time at a certain temperature, results similar

FIG. 21. SORBITE. (*Grossmann.*)
0.43% CARBON. HARDENED AND THEN TEMPERED AT 690°C. ×1,000.

to those obtainable in a shorter time at a higher temperature, it is expedient in works practice to adopt the latter method. In a steel which has been rapidly air-cooled, the pearlite may be of so fine a structure as to resemble sorbite. Such a structural condition is often described as sorbitic. A steel which consists of or contains sorbitic pearlite, as a rule, is surpassed, for constructional purposes, by the same steel which has been hardened and tempered so as to have a true sorbite structure. Thus hardening, followed by tempering at temperatures only just short of Ac_1 are often specified for constructional steels. The slight increase in hardness which may occur as the

result of low-temperature tempering of a hardened steel is due to the fact that a small quantity of unaltered austenite may have been retained by quenching. The effect of tempering such a steel at 230–320° C. is to cause this small residue of austenite to do what most of this constituent did during the quenching operation, viz. :—transform to martensite. When this has happened, an increase in hardness of the mass is to be expected. There may also be a small decrease in toughness as revealed by the impact test (Fig. 19).

The full range of tempering temperatures may for the purpose of simplification be arbitrarily divided into three sections, as follows :—

(1) 150–200° C.—The chief purpose is to release internal stresses without sacrificing hardness. This treatment is known as stabilizing and should be carried out immediately after quenching, the stresses produced by which may produce cracking in a very short time, unless relieved. Hardness and resistance to wear are the properties which it is desired to retain. The operation is usually carried out in a heated oil bath. Although not necessarily removed, internal stresses are more uniformly distributed by low temperature tempering.

(2) 200–300° C.—A distinct fall in hardness occurs in this range, but there is no gain in toughness. The martensitic structure is modified and becomes increasingly dark-etching. As already pointed out, a precipitation of highly-dispersed fine particles of carbide in a ground mass of ferrite takes the place of the lighter-etching acicular martensite. This form is granular, not lamellar like the nodular form of troostite, produced in a steel which has not been fully hardened. The Brinell hardness in this range decreases with increase in temperature.
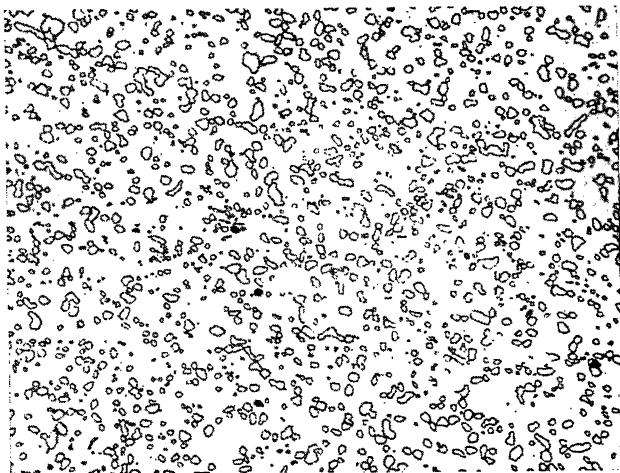
FIG. 22. SPHEROIDIZED STEEL. (0·7% CARBON.) ×500.
AFTER 30 HOURS AT 704° C., BRINELL HARDNESS 133.
(*R. E. Orton and W. F. Carter.*)*

(3) 300–550°C.—The fall in hardness continues. The increase in toughness proceeds. This softer, tougher condition is more ductile and has lower strength and elasticity. The precipitated carbide becomes recognizably spheroidized.

Being larger, the spheroids are further apart and the ferrite cushion has greater opportunity to yield along the freed planes of slip. The structure is known as sorbite. Its etching behaviour is different from that of troostite, not darkening to the same extent. Constructional steels, required to resist
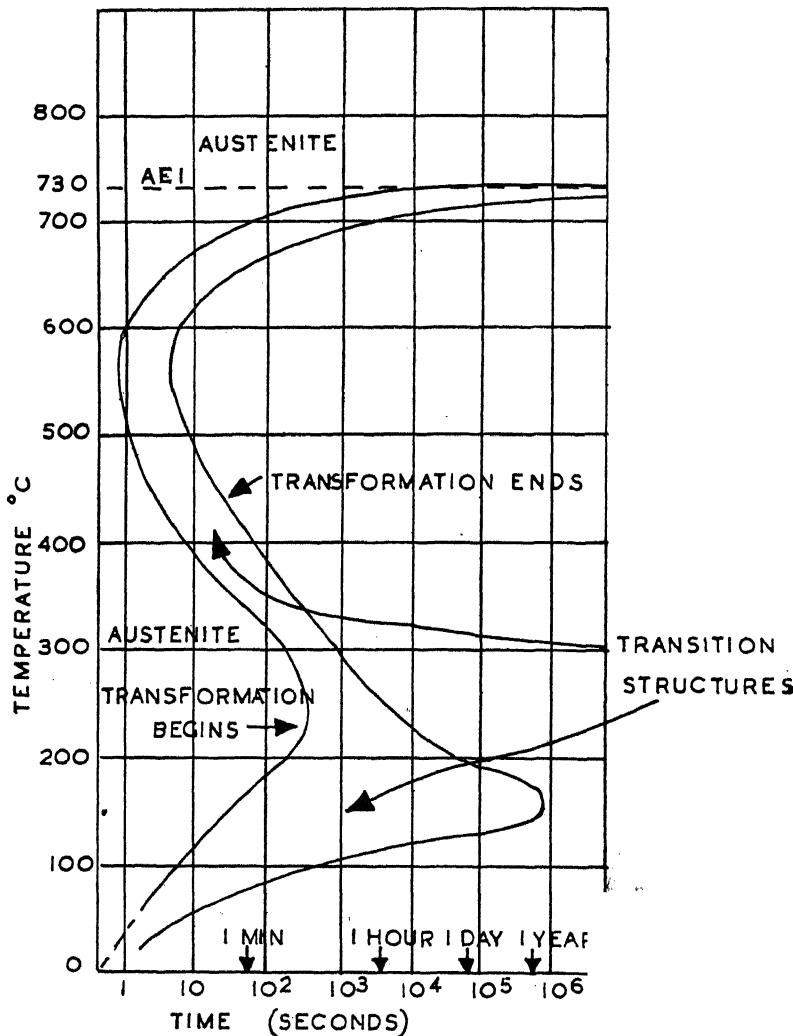
FIG. 23. S-TRANSFORMATION CURVE.

dynamic stresses, *e.g.* axles, crankshafts, large gears, etc., are treated in this temperature range, the Brinell hardness numbers varying between 220 and 350. As previously mentioned, the treatment which produces sorbite, will, if carried far enough, produce a spheroidized structure (*see* Fig. 22) *e.g.* prolonged heating of martensite at a temperature, say 675° C., just below the lowest critical temperature (viz. :—Ac_1).

By this treatment, even high-carbon steels can be made readily machineable, and workable. Steel for razor blades, for example, containing 1% carbon may, when spheroidized, be rolled to a very light gauge.

When cementite occurs in the spheroidal state, the steel will require much longer soaking at the hardening temperature, when heat-treatment has to be applied. This is because of the greater distances which the carbon atoms have to travel in order to diffuse uniformly through the ferrite cushion and convert it back to austentite.

Two-Stage Hardening ("Austempering").

It has been shown that a steel may be hardened by an acceleration in the rate of cooling from a temperature above the critical point. Quenching in cold water is one of the commonest methods of bringing about this quickened cooling rate.

The effect is not to suppress or prevent the thermal change whereby austenite is transformed but to depress this change to a temperature (say 150° C.) too low for it to be completely developed.

Instead of pearlite being the result of the transformation we get martensite. At whatever temperature the transformation begins, it will only begin after a certain interval of time and will end only after a certain additional interval of time. These two intervals of time vary with the temperature to which the steel is brought by the quenching operation. If, for example, that temperature is only a few degrees below the Ar_1 point, the first interval of time (during which austenite remains unchanged) is of comparatively long duration. The second interval, during which the transformation occurs, is of shorter duration. As the temperature reached by the steel in the quenching operation is lowered, these intervals of time shorten, particularly the first one. Between 600 and 500° C., the transformation begins after the lapse of only one second and is complete after a further lapse of about 7 to 10 seconds (in the case of eutectoid steel).

As the temperature is lowered below 500° C. the duration of the first interval increases again fairly rapidly, reaching a maximum at about 250° C., whilst that of the second interval alters very little, until, at 250° C., it also begins to increase, reaching a maximum at about 150° C.

The data now under discussion were obtained by quenching eutectoid steels in various quenching liquids held at temperatures which remained constant for the period required to complete the transformation (*see* Fig. 23).

After holding the steels for pre-determined lengths of time at these temperatures, they were finally quenched out in water at atmospheric temperature, their microstructures being then examined. At the highest temperatures (705° C.) *i.e.* just below Ar_1 the result of holding the steel for a minute or so is nil, water-quenching after this interval resulting in the same martensite structure as if the steel had been quenched without holding.

If quenched in water after the lapse of several minutes, the structure

consisted of martensite plus a small quantity of pearlite, indicating that the transformation had begun.

Quenching after a longer interval resulted in a structure in which the proportion of pearlite had increased. Only when quenched after an interval of over an hour was a structure obtained consisting entirely of pearlite, the martensite having now completely disappeared. A repetition of these holding-at-constant-temperature periods, at temperatures lower than 705° C., followed in every case by quenching, showed that the pearlite (in the pearlite + martensite structures obtained by quenching after the beginning and before the end of the transformation) was finer, *i.e.* less easily resolved at a given magnification. The same was true of the all-pearlite structures obtained by quenching after the transformation was complete. Such pearlite could be described as sorbitic pearlite.

At 600° C., the constituent associated with martensite as the result of quenching before the completion of the transformation was no longer true pearlite. It had a nodular appearance of very dark-etching propensity. Research has shown that the carbon of this constituent can exceed 0·
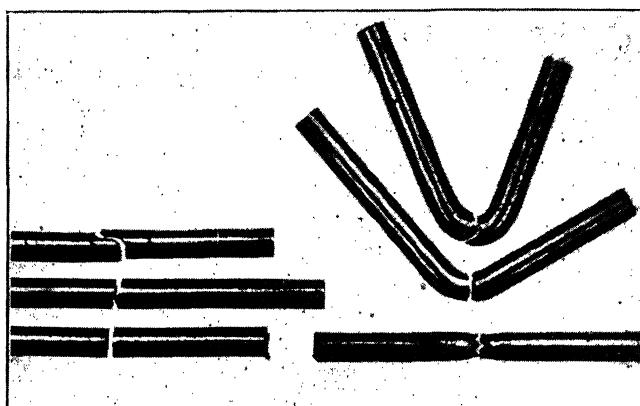
FIG. 24. COMPARISON BETWEEN "AUSTEMPERED" STEEL (*right*) AND HARDENED AND TEMPERED STEEL (*left*). (*See* Table II.)
UPPER PAIRS, SLOW BEND TESTS. MIDDLE PAIRS IMPACT TESTS.
LOWER PAIRS TENSILE TESTS.

(being the same as that of the pre-existing austenite) and although lamellar when examined under very high magnifications, the lamellæ are arranged radially to the centres of the nodules.

This constituent is known as nodular troostite and a very slight check in the cooling rate when hardening commercial steels results in its appearance.

At still lower temperatures, say 205° C., nodular troostite did not appear co-existent with martensite in specimens quenched before the end of the transformation. Instead, a dark-etching acicular constituent appeared interlaced in the martensitic pattern. This is known as acicular troostite. It was found, to the exclusion of martensite, in specimens quenched after

the transformation was complete. With these changes in structure there is a progressive increase in hardness, but only when the final quenching was carried out from holding temperatures below 150° C. on specimens in which the transformation had ended, was the maximum hardness and a true martensitic structure obtained ("white" martensite).

The dark-etching acicular troostite (the occurrence and properties of which have been closely studied by Dr. E. C. Bain) is sometimes called "black" martensite or bainite. The process of quenching in two stages to obtain this structure is known as "austempering."

Checking and holding a steel so that the austenitic structure transforms to acicular troostite before undergoing the second stage in quenching, puts a steel into a condition which compares very favourably with a similar steel quenched and tempered in the ordinary way. The former is incompletely transformed austenite and the latter tempered martensite. Without sacrificing hardness, the former has not only higher tensile strength and yield point, but also a higher percentage elongation, percentage reduction of area and impact resistance. This superiority is shown by the data given in Table II and by the illustration (Fig. 24A).
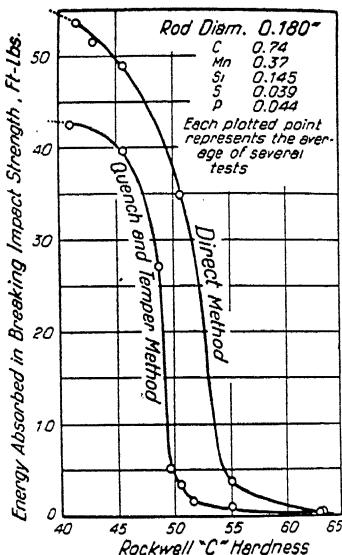

FIG. 24A. COMPARISON OF THE EFFECTS ON IMPACT VALUES, OF AUSTEMPERING AND ORDINARY HEAT-TREATMENT.

(*Davenport, Roff and Bain.*)

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TABLE II.*

Analysis.	C.	Mn.	Si.	S.	P.
	0·74	0·37	0·145	0·039	0·044

New Method.	Old Method.
Direct from Austenite	Quench to harden ; then temper.
5 minutes at 790° C.	5 minutes at 790° C.
Quench in molten lead alloy bath at 305° C.	Quench in oil at 21° C.
Hold for 15 minutes in bath.	Plunge immediately in molten lead-alloy
Quench in cold water.	bath for 30 minutes at 315° C.
	Quench in cold water.

Mechanical Properties (average of 6 tests).		Mechanical Properties (average of 6 tests).
Rockwell C Hardness	50·4	50·2
Tensile Strength, tons/in.	125·2	110·0
Yield Point, tons/in².	67·5	54·2
Elongation per cent. on 6 in.	1·9	0·3
Reduction of Area %	34·5	0·7
Impact value (foot-lbs.)	35·3	2·9

The superiority in ductility obtained by the new method is most marked in products having a Rockwell-C. hardness of about 50. It is also appreciable at other hardness values.

The new method is particularly applicable to quite thin sections and small sizes which will cool with sufficient rapidity in the heated bath.

The old-established process of *patenting* is an instance of achieving results in practice, the explanation of which is obtained later by patient research and theoretical reasoning.

The fine sorbitic pearlite structure obtainable by austempering in the high temperature region of the S transformation curve (Fig. 23) is obtained in the patenting process by passing wire as it emerges from the furnace, into a bath of molten lead maintained at the temperature which gives the desired result. The wire may be put into service in the condition thus obtained, or may be drawn down to a finer gauge.

The mechanical properties of two medium-carbon steels, after patenting and cold-drawing, are given in Table III.†

TABLE III.

MECHANICAL PROPERTIES OF MEDIUM-CARBON STEELS. HEAT-TREATED BARS QUENCHED FROM ABOVE 900° C. IN MOLTEN LEAD AT ABOUT 500° C., AS IN PATENTING, BEING SUBSEQUENTLY COLD-DRAWN.

Steel.	Original Diam. (In.)	Final Diam. (In.)	Per Cent. Reduction in Cross-sectional Area by Drawing.	Tensile Strength. Tons/ sq. in.	Elongation per cent. $(l=4\sqrt{Area})$	Reduction of area (Per cent)	Izod Value. (Ft.-lbs.)
A	0·750	0·450	64·0	74·6	9·4	27	42
B	0·9375	0·448	77·0		10·0	30	56

* Davenport, Roff and Bain. Trans. A.S.M. 1934/22].
† A. T. Adam. "Metal Treatment," p. 108.

The compositions of the two steels in Table III were as follows :—

	Steel A %	Steel B %
Carbon	0·43	0·35
Silicon	0·16	0·075
Manganese	0·63	0·81
Sulphur	0·034	0·045
Phosphorus	0·015	0·034

Harold J. Babcock* has described a practical works application of the. isothermal quenching bath (austempering). The steel was S.A.E. 4140, of which a typical analysis is carbon 0·41%, manganese 0·90%, phosphorus 0·016%, sulphur 0·031%, silicon 0·35%, chromium 1·05%, molybdenum 0·19%, used for the manufacture of shell fuzes.

In Table IV are given the data provided by various heat-treatments of 0·505 inch diameter tensile test-bars made from annealed S.A.E. 4140 steel. The first four results show that at high quenching bath temperatures, ten minutes (the time allowed for each stage of the austempering treatment) was insufficient to permit complete transformation to occur, and a mixture of high and low temperature transformation constituents was obtained in the structure. This was accompanied by low hardness and low tensile properties. Such a structure could be expected from the information provided by the S-curve of this steel.

In tests 5 and 6, the quenching bath temperatures were lowered and the test-pieces were transferred direct to a tempering bath held at temperatures higher than those of the quenching bath. The cooling rate, however, was slower than that demanded by the S-curve (6 seconds to cool from 750° to below 480° C.). The microstructures of No. 6 specimen showed that lamellar transformation-products were present, proving that the " nose " of the S-curve had been traversed. By violently agitating the quenching bath and, in tests 8 and 9, using even lower quenching temperatures, faster quenching rates were obtained. Tempering temperatures were designed to obtain Rockwell C hardness numbers of 34 to 38, the time allowed in each of the three baths being ten minutes as before. Once the isothermal quench and tempering temperatures are established for this type of steel, the operating conditions remain fixed except for slight adjustments required to meet variations in composition, grain-size, etc., as found in different batches received from the steel mills.

As applied under commercial conditions, this method lends itself to continuity of operation. Fuze parts, held in suitable basket-type racks, are carried progressively through the hardening (or " austenitizing "), the isothermal quench and the tempering baths, thence on to the cooling racks, and then into the washing and rinsing tanks, which remove the last trace of salt. They are carried through on the same fixture without any re-handling during the heat-treatment.

By using salt-baths in each of three steps, the total time of treatment was reduced to between 40 and 48 minutes per batch, compared with 1½ to 3 hours

* *The Iron Age, Feb. 10th, 1944.*

by the conventional quench and temper methods. No quench cracks of even microscopic dimensions occurred, whilst distortion was reduced to limits under 0·001 inch.

The avoidance of cracking and minimizing of distortion are due to the fact, as previously explained, that, by quenching in a bath *above* the austenite → martensite transformation, a " breathing-space " is gained of sufficient duration to enable the mass of the piece to reach *uniformity* of temperature before proceeding with the cooling (this is shown by the S-curve, Fig. 23).

No martensite has yet formed ; the structure is still uniformly austenitic. On removal from the bath, the temperature of the steel will fall to that of the transformation to martensite, with its attendant expansion. The ·deforming stresses imposed by this volume-change, under the greater quenching severity inseparable from conventional hardening practice, cause more distortion in those parts which have not yet begun the transformation, since they are at relatively much higher temperatures than the parts already transformed, and very plastic.

In other words, a greater disparity of temperature exists between transformed (martensitic) and untransformed (austenitic) parts, than exists in a piece, previously cooled in a salt bath to an isothermal condition, before undergoing further cooling to the martensite change.

In the latter case, martensite forms more uniformly at a slow rate of cooling throughout the cross-section and the stresses imposed on the still unchanged austenite are cushioned thereby. The latter, although less plastic than it was at a higher temperature and therefore less deformable, can still accommodate the stresses without cracking.

In the process above described, in which the steel is transferred to a tempering bath, from the isothermal quenching bath, the transformation to martensite is avoided entirely, the actual transformation-product being intermediate between pearlite and martensite.

The longer pre-transformation interval of time existing at temperatures between 300 and 200° C. is taken advantage of in so-called interrupted hardening processes.

Time-Quenching (Interrupted Hardening).

In this process the steel tool is quenched from the hardening temperature in water, withdrawn therefrom when still a few hundred degrees above room temperature and then allowed to finish cooling in air or oil. The drastic acceleration of cooling rate obtained by water-quenching over the first part of the S-curve enables it to escape the softening caused by premature formation of nodular troostite.

The penalty of cracking, often incurred by water-quenching, is avoided by allowing the transformation with its accompanying volume change, to occur at a slower rate. The method aims at the safety of oil-quenching, and the efficiency of water-quenching.

An alternative method of escaping the first hump of the S-curve and of permitting the austenite → martensite change to take place more slowly and more safely is to quench in water to a temperature just above the martensite-formation temperature, transfer to a furnace or liquid bath at that temperature, holding for a definite time until the change has occurred. Subsequent cooling may be slow or fast, depending on the size and shape of the article.

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TABLE IV.*

TENSILE PROPERTIES OF LOW-ALLOY STEEL (S.A.E. 4140) † AFTER HEAT-TREATMENT BY ISOTHERMAL TRANSFORMATION METHOD (" AUSTEMPERING "). TENSILE TEST BARS 0·505 IN. DIAM.

Test No.	Bath Temperature, ° C.			Hardness.		Yield Pt. Tons/sq. in.	Tensile Strength. Tons/sq. in.	Elongation. %	Reduction in Area. %
	Austenitiz-ing.	Quenching.	Tempering.	Rockwell C.	Brinell.				
1	843	621	—	23	241	35·9	50·0	24	61
2	843	565	—	26·5	264	48·2	63·4	18·5	26·5
3	843	510	—	27	265·5	48·6	72·3	14	23
4	843	468	—	20·8	265	44·9	64·7	15	30
5	843	398	580	24	248	40·4	53·5	21	58
6	843	371	565	27	265·5	47·0	56·7	19	59·5
7‡	843	371‡	510	36	341	62·5	71·4	14	50·5
8‡	843	343‡	537	34	321	68·6	75·9	17	53
9‡	843	330‡	481	38	363	84·4	89·8	14	46·9

* H. J. Babcock. "The Iron Age," February 10, 1944. † C. 0·35-0·45, Cr. 0·80-1·10, Mo. 0·15-0·25.
‡ Bath violently agitated

Chapter 2

HEAT TREATMENT OF STEEL

PRACTICAL OPERATIONS

Great care and practical experience are required for the successful heat-treatment of steel in its various branches. The hardening operation should result, with a given steel, in as deep a penetration of the hardening effect as that steel allows and the regions thus penetrated should have been transformed as completely as possible to martensite, avoiding grain-growth, decarburization, distortion and cracking. Steel becomes soft at furnace temperatures and if there is any danger of loss of shape by sagging, adequate supports must be provided to prevent such sagging while the articles are in the furnace.

Straight carbon steels are still extensively used even though inferior to alloy steels as regards resistance to distortion, depth of penetration of hardening effect and length of life in service. For straight carbon hard steel tools the range of carbon content is from 0·55 to 1·40%.

There is a carbon content to suit each type of tool and the use to which it is to be applied. For shock-resistance, *e.g.* pneumatic snaps and chisels, swaging dies, etc., a certain degree of toughness is required and this is possessed by steels containing less carbon than all-pearlite (eutectoid) steels. Such steels are called hypo-eutectoid steels. For wear-resistance and retention of a cutting edge, *e.g.* press blanking tools and shear blades, hardness is of greater importance than toughness and the carbon may be higher than that of eutectoid steel. Such steels are called hyper-eutectoid steels and, in their constitution as hardened, free cementite is found, in addition to martensite. They are more fragile in the hardened or unhardened state than hypo-eutectoid steels.

Straight carbon steel cutting tools, in the hardened condition, become softened in service, if raised to too high a tempering temperature as the result of the heat generated by friction. They will, therefore, fail, if the cutting speed, or depth of cut, exceed certain limits. Certain alloy-steels, such as high-speed steels, etc., do not become tempered when these limits are greatly exceeded. Nevertheless, much useful work is done under conditions below such limits, in which cases straight carbon steel tools may be used, with a considerable gain, as regards cost. They may also be put to many other uses, *e.g.* blacksmiths' tools, files, hammers, rock drills, etc.

Steel manufacturers often supply the various grades and qualities of steel in the form of cogged bars, which have been produced from the cast ingot by hammering or rolling. Many tools have to be shaped by forging cogged bars of suitable section.

Forging.

The temperature of the steel, at the moment when the forging or rolling operation is completed, is known as the finishing temperature.

The temperature at which the operation began, must be high enough to allow the finishing temperature to be as requisite without being so high as to cause overheating. (*See* Table V and Fig. 27.)

Cooling after forging must be uniform and the article must be protected from moisture or cold air. Air-hardening steels should be plunged into dry ashes, lime or mica dust in order to retard the rate of cooling. Annealing should follow forging as soon as possible in all cases where machining is required.
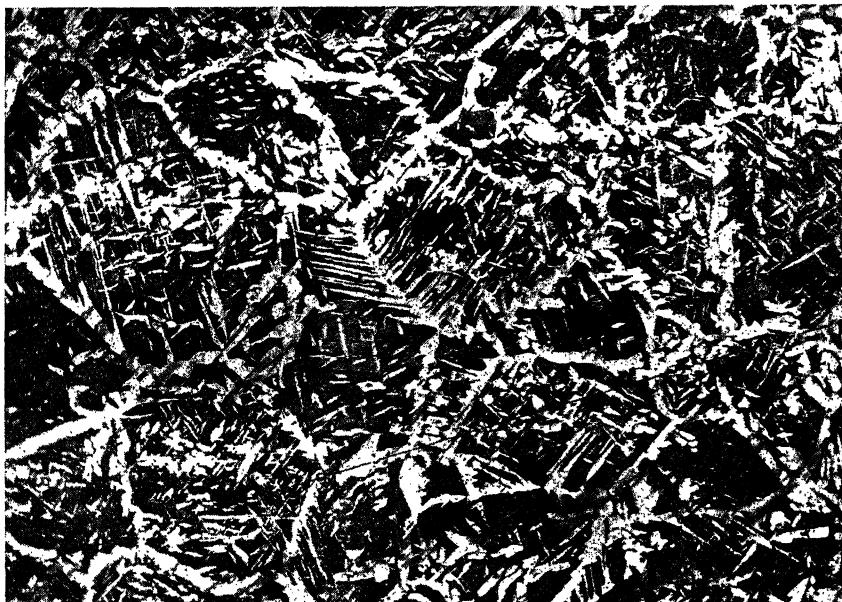

FIG. 25. OVERHEATED MEDIUM-CARBON STEEL (WIDMANNSTÄTTEN STRUCTURE). ×150. SWIVEL AXLE AS FORGED.

TABLE V.
FORGING TEMPERATURES FOR PLAIN CARBON STEELS.

Carbon %.	Max. Forging Temperatures, °C.	
	Thick Sections.	*Thin Sections.*
0·1	1315	1290
0·3	1290	1230
0·5	1260	1205
0·7	1220	—
0·9	1175	—
1·1	1135	—
1·5	1040	—

Finishing temperature should, in no case, be lower than 980° C.

Preheating.

Before charging into the furnace in which articles are to be heated for forging, annealing, normalizing or hardening, they should be preheated in a separate furnace held at a lower temperature. Two or more furnaces at successively higher temperatures may be necessary.

This lessens the risk of cracking or distortion which often occur as the result of stresses set up by differential expansion when steel is charged directly into the chamber of a furnace at the higher temperature. Heating through the lower temperature range up to 450° to 500° C. is the most critical. Slow rates of heating through this range should be favoured. One of the greatest factors in heating steel for any purpose is to bring the mass *evenly* to the required temperature. Starting with the furnace cold will achieve this and is often practised.

It is usually desirable to preheat forgings before heating to the hardening temperature, especially if they are of complicated design. Scale should be removed before hardening as it not only encourages decarburization, but, being a poor thermal conductor, it interferes with the uniform rate of abstraction of heat from the steel.

Normalizing.

When a steel is heated to just above the critical point (Ac$_3$), whatever the grain-size may have been, it can be refined and if the steel is not held at the refining temperature too long, a finely-crystalline structure, with
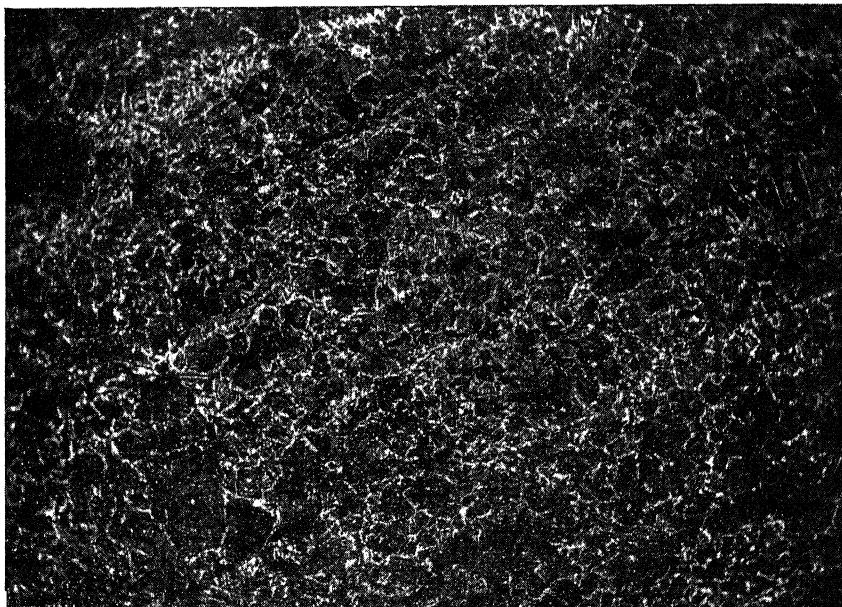
CTURE OF SAME STEEL AS IN FIG. 25, REFINEI
BY NORMALIZING.   ×150.

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superior properties, is obtained after cooling in air. If there were internal stresses they will have been relieved. If the steel was hard it will now be soft. If the steel had the unfavourable Widmannstätten structure (Fig. 25) possessed by steel castings or forged steels which have been strongly over-heated, this also will have given place to the normal fine-grained structure (Fig. 26) which a reliable steel is expected to possess. If the finishing temperature of hot worked steel has been so low as to leave it in a state of strain and structural distortion, these effects will have been removed. Articles should be heated uniformly to the normalizing temperature, allowed to become uniformly hot at that temperature and finally taken out to cool at a uniform rate in air. This uniformity in rates of heating and cooling is favoured by careful disposition of articles in the furnace and after taking them out. A single (uncovered) layer in both cases will give the best results.
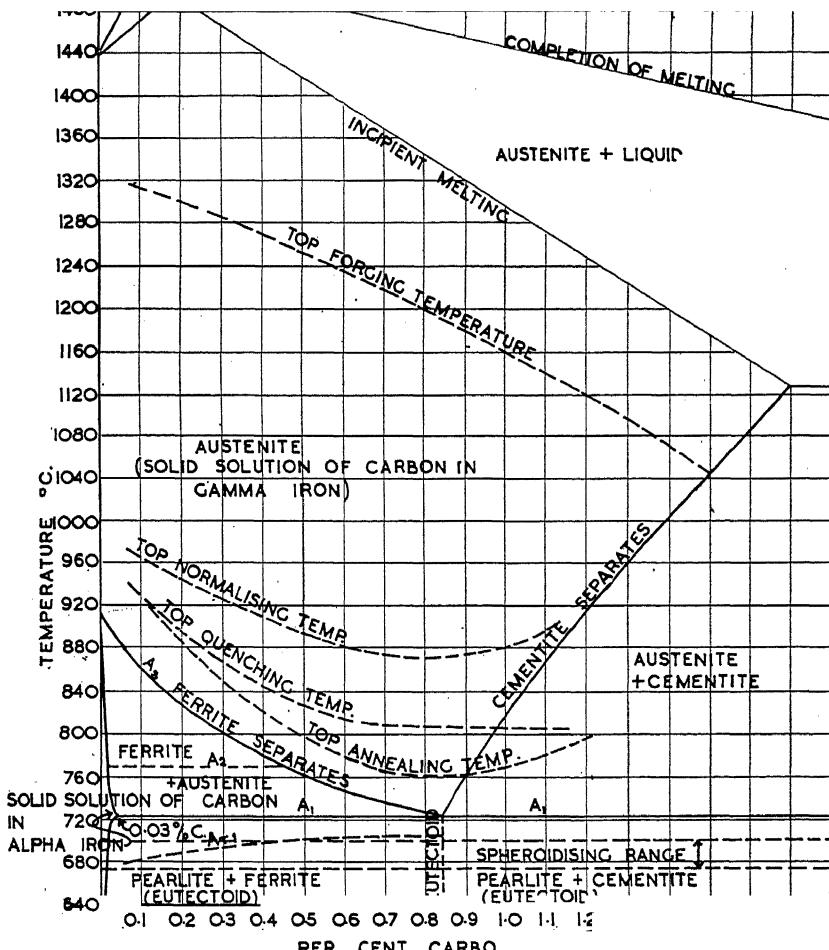

FIG. 27. IRON-CARBON EQUILIBRIUM

Annealing.

Full Annealing.

This operation is desirable in the case of tool steel after forging or rolling. The structure which is suitable for machining (or hardening) is thus obtained, a slow rate of cooling being essential, also protection from oxidation. The articles are often packed in boxes or pipes, heated to about 55° C. above the upper critical temperature (Ac_3) and held thereat for a period usually not less than one hour per inch thickness of section of the heaviest object being treated. Cooling with the furnace follows, the material only being taken therefrom when cold.

An alternative to this procedure is to allow the furnace to cool to some definite temperature below the lower critical point (Ar_1). The furnace is held at this temperature long enough to allow annealing to be complete. Removal of the containers can then be effected, followed by cooling in any convenient manner. This method has the advantage of economizing in time and fuel.

For normalizing and annealing temperatures (upper limits) *see* Fig. 27.

Process Annealing.

This operation serves (a), to soften material previously hardened ; (b) to remove internal stresses (and other undesirable effects) of operations such as welding, hard surfacing, machining, cold rolling, coining, stamping, bending, drawing, spinning, forming, punching and shearing ; (c) to restore malleability and/or ductility to cold-worked materials in order to render them amenable to subsequent mechanical treatment. The temperatures generally used are from 550° to 650° C.

The procedure is as follows :—Pack the steel in suitable containers, filling in all intervening spaces, if necessary, with the protective material most suitable in the particular circumstances. The container is fitted with a lid, charged into the furnace, held at the annealing temperature (usually just below Ac_1) for the required time and then withdrawn.

Any heating operation will be shortened by having the furnace hotter than the actual temperature appropriate to the steel and to the purpose of the operation. In this way the heat is forced more quickly to the centre of the steel, but this practice cannot be described as other than injurious, encouraging grain-growth at the surface, weakening of sharp corners and decarburization.

Gradual heating of large masses of steel, especially if of uneven section, allows thin and thick parts to heat up at nearly equal rates, thus avoiding the thermal stresses, set up by premature expansion of the thin parts when heated in advance of the thick parts.

Objects should be so disposed in the furnace as to allow free circulation of heat all round them. More uniform circulation of heat and more uniform temperatures are obtained by raising the articles above the furnace floor on some kind of support, such as firebricks.

The rate of cooling, after annealing, depends on the carbon content. Slower rates are necessary as the carbon increases. Steels containing more

than 0·5% carbon should be allowed to cool slowly either in the furnace or in a bed of non-conducting material outside the furnace. They may be removed when they are black, not before.

If the annealing temperature exceeds the critical point (Ac₃) to a serious extent, grain-growth will occur and a very coarsely crystalline structure will be obtained. This condition can be remedied by normalizing. Both ferrite and pearlite grains which are large and the pearlite which is coarsely lamellar, before normalizing, are refined by that operation.

Steel which has been rendered coarsely-crystalline by overheating, is often dangerously embrittled, as shown by the following figures obtained from a 0·4% carbon steel.*

<div align="center">TABLE VI.</div>

Treatment.	Heated to 1000° C. and air-cooled.	Heated to 820° C. and air-cooled.
Tensile strength, tons/sq. in.	47	41
Elastic limit, tons/sq. in.	26	22·5
Elongation % on 2 inches	7	23
Reduction of area, % .. /. ..	3	37

" Banded " Structure Control.

Owing to the inevitable segregation which occurs in alloy steels during solidification of the ingots, the composition across the section is not uniform. Carbon diffuses with comparative ease at high temperatures, but elements such as manganese, nickel, chromium, etc., do not. Therefore heat-treatment does not produce austenite of uniform composition, with the result that it does not transform uniformly. Ferrite, in hypo-eutectoid alloy steels, separates in bands under some conditions of transformation and as a network under other conditions.

When 3312 (S.A.E. No.) steel (carbon 0·1%, nickel 3·5%, chromium 1·6% approximate composition) is austenitized at 870° C. and is then transformed at 650° C., the ferrite separates in distinct bands. When transformed at 620° C. the banding is less marked, and when the transformation occurs at 595° C., there is very little indication of banding at all.

It is suggested that this may be explained by the relative rates (a) of formation of nuclei, and (b) of grain-growth in the ferrite and pearlite respectively at the various transformation temperatures. If the pearlite forms soon after the separation of the ferrite, banding does not occur. If there is a considerable lapse of time before pearlite forms, after the ferrite has separated, carbon can diffuse from areas rich in elements such as phosphorus, silicon, nickel and aluminium, with which it does not form carbides, to areas rich in elements such an manganese, chromium, tungsten, molybdenum, etc., with which it does.

Thus, the ferrite bands are formed at high transformation temperatures, 650 and 620° C., where the spread between the ferrite line and the line for the beginning of the pearlite formation is large, but not at 595° C., where the

* R. Whitfield. Inst. of Fuel. 19.1.1944.

<div align="center">45</div>

two lines are relatively close together. Therefore, the faster the rate of cooling through the high temperature region, the less opportunity is afforded for separation of ferrite into bands. This faster rate is achieved in production annealing of low carbon alloy steels, by cooling them in air from the austenitizing temperature to that approximating the transformation region.

Steels with a coarsely-crystalline structure require a longer time to normalize for grain-refinement than those which are already finely crystalline.

To soften air-hardening steels and remove internal stresses, they are given a prolonged soaking at about 675° C.

If a certain amount of scale formation, distortion and internal stress are permissible, the annealing operation may be expedited by charging the articles into the furnace without the protection of a container, thus economizing in time, labour and fuel. A certain amount of warping may occur as the result of annealing. Warped articles may be straightened cold but this should be followed by re-annealing to relieve the stresses set up by the straightening operation. It may often be advantageous to retard the rate of cooling, by burying the heated steel in dry lime or sand, wet ashes or sawdust, arranged in a heap, or preferably in a brick-lined pit.

Steel that has been cold-worked (*e.g.* sheet metal) is annealed in various ways in order to soften it and render it amenable to further cold working operations, bending, forming, etc.

Pack annealing consists in packing a large number of sheets on a substantial base of iron, placing a cover over the pack—the lower rim of the cover resting in sand, so as to exclude air—and heating in a furnace to the annealing temperature—which is usually below the Ac_1 temperature. The period varies from a few hours to several days, followed by slow cooling, box and furnace being allowed to cool until they reach atmospheric temperature. The pearlite is usually " divorced " by this process.

" Divorcing " means the agglomeration of the cementite lamellæ in each grain of pearlite into a single individual mass, which occupies a smaller volume than the pearlite grain from which it has been divorced.

This phenomenon resembles that of spheroidizing.

Spheroidizing.

This is effected by *prolonged* heating of iron-carbon alloys at a temperature in the neighbourhood of, but generally slightly below, the lower critical temperature range and is usually followed by relatively slow cooling. It is most effective when the steel is in the martensitic condition at the beginning of the operation.

In the case of small objects of high carbon steels the spheroidizing result is achieved more rapidly by prolonged heating at temperatures alternately within and slightly below the critical temperature range. (*See* Fig. 27.)

The object of the treatment is to produce a spheroidized condition of the cementite, thus improving ductility.

A term largely used in the U.S.A. to denote the holding of a steel at a temperature above the critical point, in order to allow the structure to

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transform partly or wholly to austenite, is " austenitize." Payson* states that steels can be annealed most efficiently if certain simple rules are observed, viz. :—

(1) The higher the austenitizing temperature, the greater is the tendency for the structure of the annealed steel to be lamellar, whereas the closer the austenitizing temperature is to the transformation temperature, the greater is the tendency for the structure to be spheroidal.

(2) To develop the softest condition in steel, austenitize at a temperature usually less than 55° C. above the critical point and transform at a temperature usually less than 55° C. below the critical point.

(3) Since the time for complete transformation at temperatures less than 55° C. below the critical may be very long, allow most of the transformation to take place at the higher temperature, where a soft product is formed, and finish the transformation at a lower temperature, where the time for completion of transformation is short.

(4) After the steel has been austenitized, cool as rapidly as is feasible to the transformation temperature, to decrease the total time of the annealing operation.

(5) After the steel has been completely transformed at a temperature which produces the desired microstructure and hardness, cool as rapidly as is feasible in order to decrease the total time of the operation.

(6) To assure a minimum of lamellar pearlite in the structure of annealed 0·70 to 0·90 carbon tool steel and other low alloy medium carbon steels, preheat the steel for several hours at a temperature about 28° C. below the critical before the steel is austenitized, then austenitize and transform as usual.

(7) To obtain minimum hardness in annealed hyper-eutectoid alloy tool steels, heat for a long time, about 10 to 15 hours, at the austenitizing temperature and transform as usual.

Hardening.

Too much care cannot be taken to obtain uniform heating of the steel throughout, as teeth or projections are appreciably weakened as the result of irregular heating or of over-heating. Steel should never be heated quickly in a furnace which is much hotter than the correct hardening temperature. The surface and the corners become overheated and the latter are prone to fly off in hardening.

As in the case of heating for forging, or for annealing, the steel should be gradually brought up nearly to the correct temperature and only when the whole piece is at uniform temperature throughout, should the final stage be completed, uniformly, and quickly, but not forcibly. It has recently been found by Digger and Rosenberg† that medium-carbon steels (0·5% carbon) are none the worse, as regards grain-size, when subjected to a *slow rate* of heating to a temperature above the critical point. Indeed, the finest grain-sizes were obtained with slow as compared with rapid rates of heating.

Dies, the surfaces of which are not to be ground after hardening, may be heated face downwards in a pan of powdered charcoal, ‡or protected with

* *The Iron Age. June, 1943.*
† *Res. Nat. Bur. Stand. Wash., 1942, 29, pp. 33–40.*
‡ *See also p. 63 " Protection from Oxidation."*

carbonaceous paste. This prevents decarburization during heating for hardening.

Suitable pastes are as follows :—

(1) Powdered charred leather, 2 parts.
   Flour .... 2 parts.
   Table Salt............. 1 part.
   Mix with water.

(2) Linseed or cottonseed oil mixed with powdered bone black or black lead.

The hardening temperature of a steel depends on its carbon content. Approximate hardening temperatures for steels of different carbon contents are as follows :—

<div align="center">TABLE VII.</div>

Carbon Content } %	0·6	0·7	0·8	0·9	1·0	1·1	1·2	1·3	1·4
Approx. Hardening Temperature } °C.	790	785	780	780	760	760	760	760	760

The temperatures given above are for water-quenching. · For oil quenching, they should be higher by 20° C. in each case.

In general, hardening temperatures are higher than annealing temperatures.

Unless the articles are small and of uniform section or unless packed in a container, preheating, as already indicated, should precede heating for hardening. Only sufficient time should be allowed *at* the hardening temperature to allow of uniform heat distribution and completion of the thermal transformation. Whenever re-hardening becomes necessary the articles should always be re-normalized before repeating the heating-for-hardening operation. Unduly prolonging the time (soaking) results in (1) grain-growth and (2) in cases where no protection from oxidation is provided, surface decarburization. Steels which possess a spheroidized structure require a certain amount of extra soaking to enable the carbide globules to go into solution.

No time should be lost in transferring the articles from the furnace to the quenching tank and care should be taken to avoid two or more pieces sticking together during the quenching operation.

Long slender articles should be held in a vertical position, when entering the quenching liquid, not horizontally ; they should be moved up and down in the liquid. Articles of varying section should be quenched so that the thicker section enters the liquid first. Taps, reamers, drills, etc., may be inserted into a length of gas-pipe before heating and removed therefrom, for quenching.

Depth of Penetration of Hardening Effect (Hardenability).

Broadly speaking, the average depth of effective hardening, in the case of straight carbon steels*, is about $\frac{5}{32}$ in. This is not deep enough for certain classes of tool, but for others it may be an advantage to have an unhardened core, possessing the property of toughness, whilst the encasing shell is intensely hard.

* *Water-quenched.*

The unhardened core is absent when the thickness of the section (a) only just exceeds twice the maximum depth of penetration obtained in still thicker sections (b). This is because the rate of cooling in (a) is sufficient to produce full hardening of the whole section but in (b) it is insufficient. In Fig. 28*, section (a) is $\frac{5}{16}$ in. thick and depth of penetration, by taking the $\frac{1}{16}$ in. layer at the centre, exceeds the normal $\frac{1}{8}$ in. Section (b) is $\frac{1}{2}$ in. and, with the slower rate of cooling necessitated by the larger amount of heat to be abstracted in a given time, only $\frac{1}{8}$ in. depth of penetration is realized from each face, leaving an unhardened core of $\frac{1}{4}$ in. thickness. If the articles to be hardened do not reach, or if they only just reach, the upper critical point, they will fail to harden satisfactorily. If, on the other hand, the critical temperature be seriously exceeded, a coarsely-crystalline structure of austenite will be obtained and the steel, when hardened in this condition, even should it escape the cracking or distortion engendered thereby, will be unduly brittle. Tempering will not entirely rectify the fragility induced by coarse grain. It is therefore incumbent on the hardener to obviate this trouble. Allowing the temperature to fall to the correct hardening temperature before quenching is no remedy : the coarse structure


FIG. 28. ILLUSTRATING VARIATION IN DEPTH OF HARDNESS OBTAINED WITH VARYING THICKNESS OF STEEL.
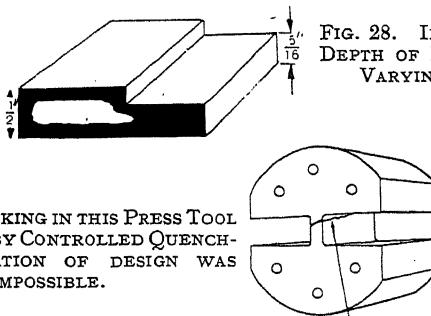
FIG. 29. CRACKING IN THIS PRESS TOOL WAS AVOIDED BY CONTROLLED QUENCHING. ALTERATION OF DESIGN WAS IMPOSSIBLE.

remains. It is not necessary to allow the articles of steel to cool down to atmospheric temperature and begin again. It will suffice to allow the temperature to fall below the critical point (say to 650° C.), then reheat to the correct hardening temperature and quench. The structure will be refined on the final reheating, which will, in effect, be an emergency normalizing operation. The soaking period at the hardening temperature—its duration appropriate to the particular steel—must always precede quenching.

The correct hardening temperature is that at which the steel will harden satisfactorily. At this temperature, the minimum distortion will be incurred. An increase in section of a steel of given composition, will require a slight increase in the quenching temperature as well as an increase in soaking time and, in certain circumstances, an increase in rate of cooling, $i.e.$ a quenching medium of greater quenching severity.

Volume Changes.

When a metal is heated it expands. When it cools it contracts back to

* J. C. Alexander. "Mech. World." 1934, 578–9.

its original size. Although the coefficient of expansion (or contraction) is not the same over the whole temperature range from freezing-point to atmospheric temperature, the temperature-expansion curve is, with most metals, a smooth one. In the case of iron, however, there is an abrupt departure from uniformity, at the temperatures of thermal transformation. At the critical point (Ar_3) when the gamma phase changes to the alpha phase on cooling, not only is there an evolution of latent heat, but there is also an expansion in volume. As the temperature falls below Ar_3 the uniform rate of contraction is resumed. In the case of steel, there is a lowering of the temperature of the volume change as the carbon increases. Rapidity of cooling, particularly by quenching, also depresses this transition temperature (Ar_3) and even when it is so severely depressed as to lead to the formation of martensite, the expansion still occurs. Moreover, its magnitude increases the higher the quenching temperature and the degree of distortion which it causes, also increases. This is illustrated in the case of a steel,* quenched from different temperatures, and carefully measured before and after hardening :—

TABLE VIII.

Quenching Temperature.	Dimensions.	Increase in diameter.	Increase in length.
760° C.	2 in. × 1 in. diam.	0·002 inch	0·001 inch
780° C.	2 in. × 1 in. diam.	0·003 inch	0·002 inch
800° C.	2 in. × 1 in. diam.	0·0035–0·004 inch	0·0025 inch

When a cold piece of steel is put into a hot furnace, the part in contact with the heated brickwork heats up more quickly than the part receiving its heat mainly by radiation and expands at a faster rate. The resultant is a volume of metal in which the more expanded part is in compression and the less expanded part in tension. Internal stresses, thus set up in the steel, remain until the temperature of the whole piece has been equalized. Conversely, during cooling, the outside layer will be cooler than the inside mass and will contract in advance of it, thus putting itself under tensile stress. Metals and alloys are not perfectly plastic ; they possess some degree of elasticity, therefore internal stresses, so long as they do not exceed the elastic limit, will disappear when the temperature is equalized. Once the elastic limit is exceeded, further thermal elongation or contraction can only occur by virtue of plastic movement. Thus different degrees of elongation will demand different degrees of elastic deformation or of plastic deformation or varying combinations of both. Plastic (permanent) deformation in one part and not in another must lead to warping or distortion. · Equalization of the different elongations produced by differential expansion is often effected partly by elastic and partly by plastic yielding. The metal is thus put under internal strains, the degree of self-straining increasing with the amount of elastic equalizing. As a rule, the ratio—elastic/plastic deformation—is greater the lower the temperature and so the internal strain will be more pronounced at low temperatures.

* *Inst. Prod. Eng. Tech. Bulletin. May, 1943.*
See also A.S.M. Trans., 1940, 28, No. 2, pp. 472–512, E. Ameen.

Steels usually have sufficient plasticity to survive the stresses created by slow rates of heating or cooling, the differential volume changes being equalized by plastic yielding. But the differential expansions, caused by the sudden introduction of the steel into a hot furnace, may produce stresses of such magnitude as to exceed the internal limit of cohesion and cause cracking. A very sudden fall in temperature may have the same result. These effects will be encountered more frequently when the metal is of such a high carbon content or in such a structural condition that it is unable to stretch, under the stresses imposed, without breaking.

This self-strained condition may be greatly aggravated, in the case of a quenched steel, when the centre of the mass not only cools and contracts more slowly than the outer layer but even expands, as it will at the temperature of the gamma to alpha transformation. The expansion resembles that which occurs when water in a pipe freezes to ice. In the case of a quenched steel, the outer layer is put under a severe tensile stress and may burst. In certain cases the differential volume changes occurring in the steel may take place in such an order that the outer layers are in compression and the inner parts in tension.

Cracking.

Cracking may not occur immediately after quenching. It may not occur until the steel is tempered, in which case the failure will take place in the early stages of that operation. Conditions favouring cracking are :—

(1) The presence of non-metallic inclusions such as manganese sulphide, or slag (*see* Fig. 33) in excess. Free cementite (especially intercrystalline) also encourages cracking. (*See* Fig. 10, Chap. 1.)

(2) Large grain-size (due to overheating).

(3) Uneven quenching.

(4) Irregularity of sectional thickness. Also sharp re-entrant angles (bad design).

1. Complete freedom from *non-metallic inclusions* is not possible in commercial steels, but by the careful selection of the raw materials used in the making of the steel, choice of best method of melting etc., steels sufficiently clean internally to behave satisfactorily in hardening, can be produced.

Cementite, in the form of an intercrystalline network, is frequently the cause of " spalling " in case-hardened steel components. The case cracks when hardened and may become a detachable outer shell.

2. *Grain-size* is of importance from several points of view and the following are some of the factors involved :—

(A) When a steel is heated to just above the upper critical point (Ac_3) and cooled in air, a grain-refinement takes place.

(B) A fine-grained steel has better mechanical properties, especially toughness, than a coarse-grained steel.

(C) Overheating causes grain growth, with resultant weakness.

(D) *Fine-grained steels* differ from coarse-grained steels as follows :—
    Fine-grained steels.
    (a) Do not coarsen readily unless considerably overheated.
    (b) Have less hardenability, *i.e.* are shallow hardening when quenched.
    (c) Have greater toughness when quenched and less liability to crack or distort.
    (d) Are less easily machined.

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Coarse-grained steels.
(a) Coarsen rather easily when reheated.
(b) Have good hardenability, *i.e.* tend to harden deeply when quenched.
(c) Are less tough and more likely to crack when quenched.
(d) Have good machining properties (alleged).

3. Uneven quenching aggravates the trouble due to mutually antagonistic volume changes and stress concentrations.

4. *Design.*—The relation of design to heat treatment is extremely important. Neglect of the phenomenon of differential volume changes introduces serious complications in the response of steels to heat treatment, and may often be the cause of failures. J. C. Alexander* gives the following instances :—In the case of pneumatic snaps, the top edges broke off completely when quenched (Fig. 30A). This was attributed to insufficient thickness of section at the top edge, in which penetration of the hardening effect was complete. Thus this part of the tool had undergone the austenite to martensite volume change (sudden expansion) before the body of the tool had reached that stage. The imposed stress overcame the resistance of the steel and fracture ensued. By increasing the thickness of metal at the top edge of the tool (Fig. 30B) the difficulty was overcome, depth of hardness penetration was even throughout and disunity of behaviour was prevented.

30A.   30B.
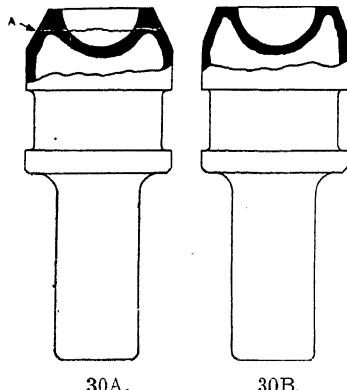
FIG. 30A (*left*) SHOWS HOW A COMPLETELY HARDENED RING CRACKED FROM THE HEAD OF A RIVETING CUP DUE TO DIFFERENTIAL CONTRACTION. FIG. 30B (*right*) SHOWS HOW THIS DIFFICULTY WAS OVERCOME BY LEAVING MORE METAL AT THE TOP OF THE TOOL. A—PLANE OF FRACTURE.

Another example was instanced in the case of a press-tool punch (Fig. 29) p. 49. When quenched, face first, cracking occurred in the corner of the narrower part, as shown. As alteration of the design was not possible, the difficulty was overcome in the quenching technique. By quenching slowly, on its side, followed by complete immersion, the heavy sections cooled and contracted at such a rate that a more even distribution of stress resulted and cracking was prevented.

* *Ibid.*

In Fig. 31 (Rollason)* methods of improving unsuitable design are illustrated. All radii should be made as large as possible, and sudden changes of section avoided, as there is a severe concentration of stress at all sharp corners, where cracking often begins. (*See* also Fig. 32.†)
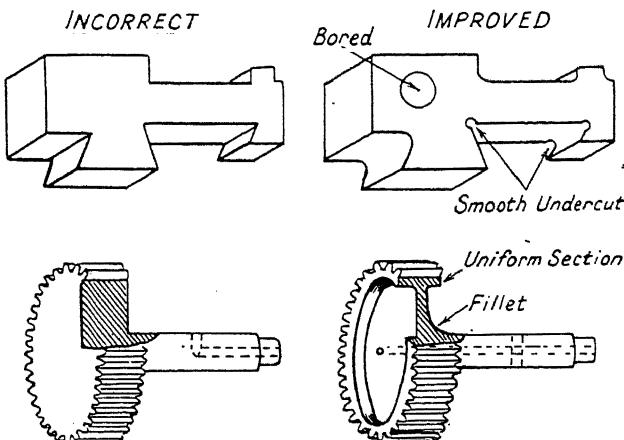

FIG. 31. THE INFLUENCE OF HEAT-TREATMENT ON DESIGN.
(*Rollason.*)

When holes and slots interfere with equalization of rates of cooling and uniformity of thermal contraction, they produce concentration of stress which encourages distortion and cracking. The plugging of holes and slots by means of a stiff paste of a mixture of asbestos, clay and graphite prior to heating gives greater uniformity to the quenching effect and obviates the formation of steam and air pockets, which cause soft spots. Teeth of cutting tools should be slightly radiused.

The practice of withdrawing direct-hardening steel from the quenching
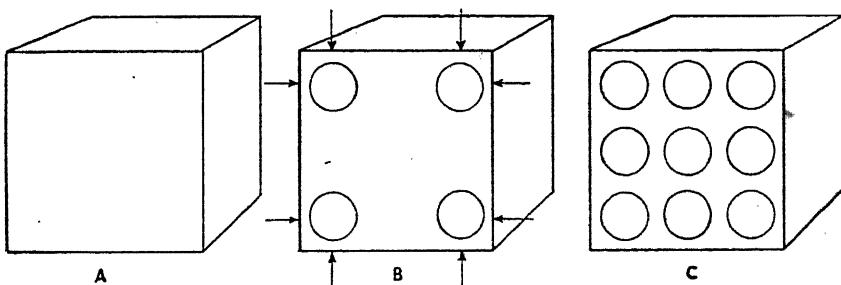
FIG. 32. (A) A SIMPLE PART SUITABLE FOR WATER QUENCHING.

(B) THE INTRODUCTION OF THE FOUR HOLES MAKES THIS PART "COMPLEX," DUE TO THE THIN SECTIONS ADJACENT TO THE THICK SECTIONS.

(C) INTRODUCTION OF MANY EQUALLY SPACED HOLES MAKES PART OF UNIFORM SECTION AND THEREFORE "SIMPLE" FROM THE POINT OF VIEW OF QUENCHING HAZARDS.

* "*Metallurgy for Engineers.*"
† W. E. Bruse, "*The Iron Age*," *June 24th, 1943, 63–67.*

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bath while it still retains enough heat to enable it to take a permanent set, allows of straightening. The structure is still austenitic and straightening by hand or even by a press is a safe procedure. If these bad features cannot be avoided it may be necessary to use an alloy steel less prone to give trouble. As the hardening effect in such a steel can be obtained with a much slower rate of cooling, the magnitude of the quenching stresses is very considerably lessened. In an air-hardening steel (*e.g.* 4% nickel, 1·3% chromium), the quenching operation, with its attendant risks, can be omitted. An air-hardening steel allows of hardening throughout the mass, even when a large mass is used. Therefore press-tools and dies of fairly complicated design should be made of an air-hardening steel*, thus permitting more sharpness at corners and greater variation of section. Some dies are so large that it is necessary to build them in sections.

In carbon steels, the normal distribution of the cooling stresses is only slightly altered if the quenching temperature is raised higher than just above the critical range. In articles of cylindrical shape, compression stresses in the surface layers and tensile stresses in the centre appear always to be derived from quenching. A heat-treated 0·7% carbon steel piston-rod, when slit longitudinally along the axis shows this in the direction of bending caused by the release of residual stresses. Each half of the piston-rod bends inwards, as the result of the outer surface layers elongating when the compression stresses residing in them are released by the slitting. The centre of the rod, released from the tensile stresses, follows a concave path, as the result of shortening. The cooling stresses gradually increase, with increase in quenching temperature, in ingot iron and in carbon steels, up to critical temperature. Quenched from temperatures above this they remain fairly constant.

The highest stresses are reached in steels with about 0·3% carbon, there being no further increase with higher carbon contents. Only from above 600° C. does *oil*-quenching produce residual stresses which are, however, considerably lower than those produced by quenching in water from the same temperature.

With increase in temperature of the water, the quenching stresses fall to very low value at the boiling point. The magnitude of the stresses is independent of the dimensions with water quenching, but with air-cooling it increases with the size of the test specimen. High cooling stresses up to about 18 tons/in² and even cooling cracks have been found in such commercial products as boiler plates, and large forged shafts. In hollow piston rods of an ocean liner engine, made from a heat-treated steel (mentioned above) tensile stresses of about 24 tons/sq. in. have been found near the surface of the bore in both the longitudinal and tangential directions. Residual stresses, up to 18 tons/sq. in., have been observed in commercial chromium-molybdenum steel tubes in the normalized condition ; the stresses at the outer surface in this case being tensile stresses, their presence increases the tendency to crack during welding.

The existence of compression stresses in the outer layer and tensile stresses in the centre can be explained by assuming that, at the moment when austenite in the outer layer transforms to martensite with accompanying expansion, the inner portion is contracting. The resultant would be an

* (*e.g.* 2% *carbon*, 12% *chromium*.)

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equalizing in which the outer shell had been checked in its endeavour to lengthen (thus being put under compression stress) and the centre checked in its endeavour to shorten (thus being put under tensile stress).

Cold-worked steels (bright-drawn bar or pressings) often show an increase in volume after quenching. The increase in temperature before quenching has enabled compression stresses to be released. These stresses are often localized, *e.g.* in the centre of a drawn rod ; their release may result in unequal expansion in different parts of the piece. This phenomenon is responsible for the ovality often found in circular rings :—free wheel components, ball races, etc. It is therefore impossible to avoid distortion when hardening components made from cold-worked metal, unless quenching presses or jigs are employed.

The liability of steels to distort in the hardening operation can be turned to advantage in some cases. For example, the 5% tungsten-steels used for wire-drawing dies can, when worn, be brought back to their former size by rehardening.

The volume-changes occurring in this operation result in the contraction of orifice back to the size it was before being enlarged by wear. This operation can be repeated whenever necessary and thus enables the die to give longer service.
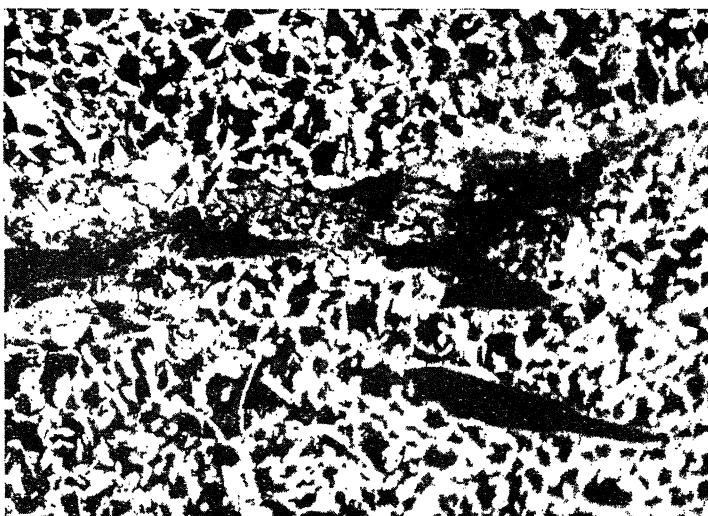
FIG. 33. NON-METALLIC INCLUSIONS IN LOW-CARBON STEEL.

There is a tendency for hardened steels to undergo growth during service.

Ageing and sub-zero treatment have been practised in order to stabilise such articles as gauges, mandrels, dies, arbors, etc.

Sub-zero treatment at $-85°C$ followed by tempering above 150°C (a cycle which is repeated about 6 times) accelerates this stabilising effect.

The growth is due to a slow transformation in residual austenite (*see* p. 19) to martensite. Sub-zero treatment, followed by tempering, hastens this change.

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Quenching Media.

A correct cooling velocity is necessary for the correct hardening of steel. The quenching medium must abstract heat from the steel throughout the maximum cross-section at a rate equal to or greater than the critical cooling velocity, if complete through-hardening is to be obtained. The quenching severity of any quenching medium is greater, the higher its thermal conductivity, its specific heat, its boiling point and latent heat of vapourization, and the lower its viscosity and temperature. For any steel in any quenching liquid, the critical cooling velocity of the steel varies with composition, grain-size, mass and initial quenching temperature.

Water.

Water-hardening steels should be quenched :—
(a) in water with forced circulation,
(b) in still brine.
Brine should be made up by adding 13-ozs. of table or rock salt to each gallon of water. The following limits should be observed :—

8 to 10% of salt by weight ;
1·006 to 1·071 Sp. Gr. ;
15° to 27° C. when in use.

Common salt raises the boiling point and lessens vaporization. Other salts have a similar effect.

Straight carbon steels are more effectively hardened by quenching in water and give better results after tempering than when oil-quenched, even though subsequently tempered at a lower temperature in the latter case. A higher hardening temperature is necessitated when the action of the quenching medium is less severe, as in the case of oil. The higher the carbon content the greater the risk of distortion or even cracking, when steels are water-quenched. The harder chromium and nickel-chromium steels are quenched in oil, the lower alloy steels, however, requiring a higher quenching temperature.

The temperature of water in the quenching tank will soon rise above the specified range, as the result of quenching a large number of articles in quick succession. Consistent hardening results will not then be possible.

Methods of circulating the water, where large numbers of articles have to be hardened, are adopted.

For large, solid tools, the temperature of the water should be about 10° C. and, for intricate articles, about 27° C. The surface of the water should be free from oil-films, which, by adhering to parts of the surface of the articles, check the rate of cooling and produce soft spots. Soap in solution has an effect opposite to that of salt and must be avoided.

Oil.

Parts liable to become warped should always be oil-quenched and should preferably be held in jigs during the operation. Warping, however, cannot be entirely prevented with articles of complicated form. Care must be taken, even with alloy steels, not unduly to exceed the normal hardening temperatures, otherwise some diminution of the toughness and abnormal retention of hardness may occur, especially after low temperature tempering.

Important properties of quenching oils are :—(1) viscosity, (2) volatility, (3) resistance to oxidation.

(1) *Viscosity.*

High viscosity means sluggish rate of renewal of fresh liquid to replace that locally heated by the hot steel. This heated layer becomes lighter and should therefore rise to the surface, allowing surrounding cold oil to replace it. Convection-currents are checked by liquid of high viscosity and encouraged by thin oils. Thin oils, however, vaporise rapidly and deteriorate. When oil is warmed, its viscosity decreases, its quenching severity is thereby increased, but it produces less acute internal stress. (*See* Fig. 34.)
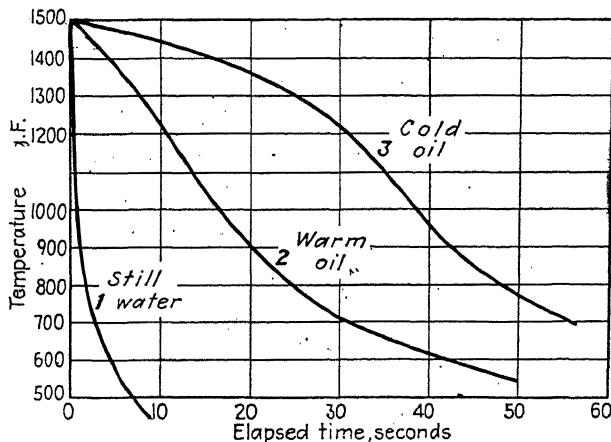
FIG. 34.   COOLING RATES OF STEEL WHEN QUENCHED IN WATER
AND OIL.

Not only does sluggish movement of the oil result from high viscosity but also deterioration, due to charring of the layer adjacent to the hot steel and checking of rate of heat transference with possible incorrect hardening of the steel.

The " drag-out " loss of thick oil is higher than that of thin oil.

(2) *Volatility.*

The lowest volatility possible, consistent with satisfactory viscosity, is desirable. High volatility may result in uneconomical working, owing to high volatilization loss, accompanied by automatic thickening as the volatile fractions forming the thinning constituents of the oil are distilled off. Soft spots and non-uniform hardening may result.

Flash-Point.—When the hot steel enters the tank, highly volatile (low flash point) oil will burn and smoke. The most serious result is uneven hardening, due to gas-formation at the commencement of the operation (non-conducting gas films). These films and pockets of vapour (especially in recessed parts) insulate the steel and check the rate of cooling.

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(3) *Resistance to Oxidation.*

All hydro-carbon oils tend to thicken and throw down a sludge. Sludge is caused by oxidation and is governed by three factors, viz. :—(1) time ; (2) temperature ; (3) surface area exposed to oxidation. Oxidation causes thickening and sludge of low thermal conductivity collects on the tank bottom.

Pieces of steel sinking into this are imperfectly hardened.

Sludge also clogs the pipes of the circulating system (where used). This hinders circulation and lowers the cooling efficiency of the system. Moreover, " drag-out " losses increase with thickening and the quenching efficiency of the oil falls off, causing the results of the hardening operation to vary.

Oil should be a light mineral oil :—

     Minimum flash-point 117° C.

     Ignition Temperature 200° C.

     Viscosity 100 Sec. Saybolt at 43° C.

     Temperature range of use, 43°–65° C.

Oil should never be agitated with compressed air and its temperature never allowed to exceed 65° C. It should be filtered or reconditioned if it becomes dirty or viscous.

The use of pure mineral oil is essential for the quenching of steel direct from molten salt baths. Such baths contain sodium and potassium salts, which saponify any fatty matter contained in oil, *e.g.* whale oil.

Experiments carried out by Stanfield on quenching tests in various media (J. I. and S. Inst. : 1939, Second Alloy Steels Report, p. 299) show that, qualitatively, when suitable observations of temperature are made during quenching, the use of 3 inch diameter steel test-pieces indicates the relative quenching capabilities of different oils for masses of moderate dimensions.

The relationship between the rate of heat abstraction by quenching medium and the surface temperature of the article being quenched does not appear to follow any simple or definite law.

A relationship similar to Newton's Law—that the rate of heat transference is proportional to the difference in temperature between the hot body and its surroundings—is at the best a very rough approximation to what really occurs, although such a law may be approached more nearly in the later stages of cooling. The conditions of cooling, at the commencement of quenching, cause a rapid rate of cooling of the skin of the article to a very shallow depth and, in certain applications, such as thin articles, this may be of importance.

The physical properties of the material being quenched appear to be equally important as the properties of the liquid, in controlling this rate of cooling of the skin at the beginning of quenching. In particular, a low thermal conductivity of the metal tends to decrease the depth of skin thus affected and to increase its rate of cooling.

It is clear also that, for quenching steel articles of moderate sizes, the skin temperature falls so rapidly through the first 200° C. to 250° C. that the characteristics of the quenching medium, at temperatures within this range do not appreciably affect the quenching operation as a whole, and the rate of cooling, conferred on the major portion of the steel mass, so far as the quenching

medium is concerned, depends essentially on those properties of the oil in contact with surfaces at temperatures below say, 500°C. to 550° C. It has been shown that the surface condition of the mass being quenched, substantially affects the rate of heat abstraction, a scaled surface giving a more rapid rate than a bright surface. This result would be expected if the scale had the effect of opposing the maintenance of a vapour film.

Alternatively, this effect may be connected with differences between turbulent and stream-line flow in the quenching medium.

The effect is of smaller magnitude for the more rapidly quenching type of oil. The temperature of the quenching oil should not be allowed to fall below or rise above a certain appropriate range of temperature.

Variations in results accrue from neglect of this precaution.

Gregory and Simons* state that other factors, all of which can be ascertained by suitable known tests, are as follows : (a) minimum carbon content, which means that there is greater likelihood that the oil will maintain its efficiency during prolonged use. Carbon is essentially unstable in many organic compounds, and soon reacts, under heat, to form a sediment ; (b) low water content, so that less water vapour is formed, and also that the too drastic and non-uniform cooling action of its presence is prevented; (c) low content of certain fatty oils, which smell unpleasantly under heat and also deteriorate and decompose. This means that, for preference, the oils should have a mineral base, but it is generally better to keep them within certain temperature limits by a cooling system.

The significance of oil tests is not all-important, however. The various tests must be regarded rather as safeguards to ensure that supplies of an oil already found suitable are fully up to standard. The most vital factor in quenching oil is its ability to maintain its properties over as long a service period as possible.

Protection from Oxidation.

Iron is a metal which, when heated in the presence of oxygen, is converted at the surface to oxide. The oxide of iron thus formed has a low vapour pressure and is compact and physically cohesive. This type of oxide layer exercises a pronounced restraint upon the continued activity of oxygen and hinders the rate of oxidation in accordance with :—(a) the degree to which oxygen will dissolve in it and (b) the speed with which oxygen will diffuse through it from its outer to its inner surface. If absolutely impervious to oxygen, as indeed a film of alumina on aluminium is, then further oxidation is entirely prevented. If permeable, then oxidation will proceed. The rate of diffusion of oxygen varies inversely as the length of path of diffusion i.e. the thickness of the oxide layer. Treated mathematically, a parabolic curve is obtained, the quantity of oxygen combining being proportional to the square root of the time of oxidation. This refers to physically perfect oxide sheaths and may be experimentally realised by continuous heating at constant temperature. A factor of considerable importance to be borne in mind, however, in determining the rate of oxidation at high temperatures, is the resistance of the oxide sheath—by virtue of its plasticity and strength —to the disintegration caused by the disparity between the co-efficients of expansion of the metal and the oxide. Thus, when iron is exposed, at high

* " *Edgar Allen News*," *July, 1944.*

temperatures, to atmospheric oxygen or other oxidizing gases (*e.g.* carbon dioxide and steam), the surface layer of metal becomes converted to oxide (scale). The longer the time of exposure and the higher the temperature, the thicker the layer of scale becomes, although, as already explained, the rate of oxidation progressively slows down. The permeability of this layer is favourable to further oxidation in iron and in carbon steels. The composition of the scale is complex. The oxides of iron form a group with complicated relationships between temperature, pressure and constitution; mutual solubility between the oxides adds a further complication.

The scale consists of two or three layers. The innermost layer, adjacent to the metal, contains less oxygen than the superposed layer (or layers). It is well-known that the innermost layer adheres strongly to the metal, but the outer layer or layers are readily detachable.

The greater the amount of scale formation the greater the loss of metal, the greater the cost of removal by shot-blasting, pickling, etc., and, in the case of steel, the greater the removal of carbon from the surface layers of the metal (decarburization). Direct impingement of the gaseous products of combustion on the surface of the metal is very dangerous. These gases are at a very high temperature, to which the skin of the metal can easily be raised, resulting not only in heavy scaling and severe decarburization but also in intercrystalline oxidation of the metal (" burning ").

The encasing shell of scale on experimentally-heated wires is prone to form as a hollow, swollen tube surrounding the wire. The surface is frequently contorted into grotesque shapes, very often with crests and protuberances of reddish ferric oxide, depending somewhat on the temperature of oxidation. When cold, the shape of these suggests that, during either
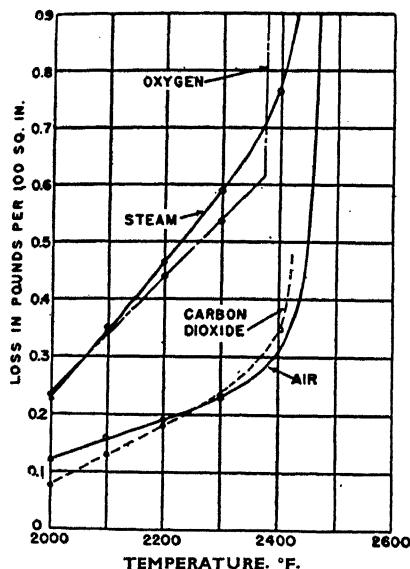
Fig. 35. Effect of Variation of Temperature on the Scaling of 0·10–0·20% C. Steel (S.A.E. 1015).

oxidation or cooling, a gas liberation occurred, puffing the oxide shell out and sometimes bursting it, leaving small crater-like mouths. In almost every case the oxide, when cold, was loose and of considerably larger internal dimensions than that of the wire which it sheathed. This behaviour is apparently closely related to the concentration of oxygen in the atmosphere, for Armco iron generally yielded smoother and more symmetrical layers in air than in oxygen, while an extremely perfect coating was formed during oxidation at a much reduced pressure (10 mm) followed by cooling *in vacuo*. The author has seen smooth, dense oxide sheaths, 1 in. in diameter, surrounding a residual core of steel ⅛ in. in diameter, as the result of leaving a steel rod in an annealing furnace for a very long period.

Reactions, other than oxidation in its simplest sense, appear to be in operation and, after the formation of the early layers, erratic departures from the quadratic proportionality occur. Armco iron oxidized at a consistently faster rate and formed coatings which were, in general, smoother than those of electrolytic iron.

The influence of increase in temperature upon scale-formation on mild steel (0·13 to 0·18% carbon) is shown in Fig. 35. In addition to air, the action of carbon dioxide, steam and oxygen are illustrated. It is interesting to observe that, at certain temperatures, steam can be the most effective of all four gases mentioned and that carbon dioxide above 2250° F. (1230° C.) is more effective than air.

To protect steels from oxidation when in a hot furnace, they are often surrounded by some material which will use up the oxygen in any air which may be in, or gain access to, the containers. The articles are packed in the protective substance, which may be " spent " carburizing material, ground plumbago pots (" brasque "), cast iron swarf, pitch-coke, etc. A little resin placed in the container is effectual in preventing decarburization and renders other filling-in material unnecessary, except when required to prevent sagging.

Cast iron swarf should never be used more than once, as it becomes converted to iron scale, which not only fails to protect the steel but has a positive decarburizing action of its own. The other materials may be used repeatedly, sieving occasionally to remove fine dust.

Carburizing materials may be used in the temperature range 590° C. to 1025° C. At temperatures above the critical point, they have a surface hardening effect which is not usually detrimental, except when maximum toughness is required or when the design of a tool is such that carburizing of the edge may cause it to chip or break. It may be quite beneficial in the case of drawing or forming dies, lamination dies, etc. High speed cutting tools and steel-tools used for hot-working operations should not be packed in carburizing compound during heat-treatment. In cases where it is necessary only to protect a die cavity or a working edge, especially if time is important the die or tool is heated up to 600° C. and after sprinkling with boric acid, is charged into the hardening furnace, for subsequent quenching. The boric acid melts and covers the surface to which it is applied with a protective coating. This coating is removed by the quenching operation, when water is used, leaving a clean hard bright surface. This " particular " treatment should not be used when temperatures above 1010° C. are used, nor when

oil is the quenching medium. Proprietary materials for surface protection of steels during heating operations are obtainable from firms specializing in steel manufacture and treatment. The steels are cleaned, heated to a suitable temperature and then coated with the protective material before heating to the hardening temperature.

No special precautions are necessary in the case of tools which have to be ground to size *after* hardening. Tools which may not be so treated or on which fine cutting edges are required should be heated in furnaces designed to have a non-oxidizing atmosphere.

Sometimes it is desirable to confine the hardening effect to one part or parts of the article, leaving the remainder unhardened and machineable. The face of a die may be water flushed, leaving the under part soft. Holes in dies and constructional parts (sometimes requiring to have an internal screw thread) should be filled in with a suitable clay, fire cement or asbestos. Chisels and similar tools, only the ends of which are to be hardened, may be tempered in one and the same operation by allowing the heat in the unquenched part to be conducted back to the hardened end. The correct stage of tempering is judged by the colour of the oxide film which forms on the previously brightened face.

The Use of Molten Baths.

There is always the possibility, in heat-treatment furnaces, that the work itself may not be at the temperature indicated by the pyrometer. When a bath of molten salts is used, not only may the required temperature be accurately obtained, but uniformity of temperature throughout the bath is not difficult to secure, owing to the distribution of heat by conduction and convection. The temperature of the bath is lowered whenever cold steel articles are introduced and time must be allowed for (*a*) its restoration to the required value, and (*b*) the soaking of the steel, the length of this additional period depending on the size and composition of the articles undergoing treatment. For cutting tools, $1\frac{3}{4}$ minutes per inch of cross-section is a fair approximation to the time usually required to attain the temperature of the bath throughout the tool.

The bath affords protection from oxidation during the heating period and a thin film of salt which adheres to the work when it is removed, continues this protection during transference from salt bath to quenching tank. As a result, not only is the formation of surface scale avoided, but decarburization also, a matter of considerable importance when treating such articles as files, small taps, etc.

If decarburization occurs when using certain salts, the addition of sodium cyanide will prevent it. A bath of low cyanide content (under 20%) can decarburize high carbon steel if given enough time. On this account, when heat-treating steels of 0·8% carbon and over, the ordinary cyaniding bath (50% cyanide, 50% sodium carbonate) should be used. Only cyanide additions are necessary periodically to keep this bath up to its original level and up to working strength. The carbon content of the steel will not be caused to exceed 0·9%, at temperatures up to 950° C., which is the maximum working temperature for such a bath. Steels of lower carbon content than 0·8% would become carburized in this 50/50 bath and therefore neutral baths must be used (*e.g.* Cassel Heat Treatment Salt). The latter are also

suitable for normalizing and hardening parts which have been machined after carburization. When surrounded by molten salt at a uniform temperature, the rate of heating of the work will be uniform. In the case of articles of uniform section, such as gudgeon pins, shackle bolts, nuts, free-wheel components, etc., no special care need be taken to preheat at a lower temperature, even though the rate of heat transference by conduction (as distinct from radiation in a muffle furnace) is very rapid. In parts having inequality of section, however, the sudden change when introduced into the highly-heated bath will result in the thinner sections expanding in advance of the thicker sections and producing distortion. By (a) preheating such parts gradually to 300° C. ; (b) immersing them in the bath at a lower temperature ; (c) raising the temperature to that required for hardening; uniform rate of heating and expansion are obtained and distortion avoided. When the normal expansion of a steel article is countered by sudden contraction at the temperature of the thermal transformation, the stresses set up may cause distortion even in articles of uniform section if the rate of heating is excessive. Molten baths, by reason of the suddenness with which they heat up the outer layers of steel articles immersed in them, are prone to aggravate this trouble. To avoid the simultaneous occurrence of expansion and contraction in the same article, it should be brought to a uniform temperature just below the transition point through which it may then be allowed to pass, so that all parts undergo the contraction, caused by the constitutional change, at the same time.

The following rules may be given as representing the best practice to avoid distortion, viz. :—

(1) *For Case-hardening Steels.*
(a) Immerse in bath at 700° C. to 760° C. and soak at 760° C. until the
. piece is at uniform temperature.
(b) Raise the temperature to 890° C. to 900° C. and again soak.
(c) Raise to carburizing temperature (950° C.).
(2) *For Direct-hardening Steels.*
(a) Preheat to 550° C. Transfer to salt bath *at* 550° C.
(b) Raise bath slowly to the hardening temperature and quench without
further soaking.

As correctness of quenching temperature of steel is of great importance in avoiding cracking, heating in molten salt baths has the advantage of giving ease of temperature attainment and accuracy of control. These factors are advantageous, also, when molten salts are used for tempering, in which operation, temperature fluctuations or inaccuracy may be fatal to the task of meeting specifications.

H. J. Babcock* states that the commercial application of the isothermal procedure in the quenching of steel, requires, for its success, a bath capable of absorbing and dissipating a large quantity of heat without progressively increasing in temperature. Molten salt baths, electrically heated by immersed electrodes, with automatic temperature control and no external supply of heat, are used for this kind of treatment. Any increase in the temperature of the bath is prevented by increasing the rate of flow of cooling-air through an air-jacket surrounding the bath. Lueg and Pomp† have shown that,

* " The Iron Age." February 10, 1944, pp. 62–69.
† " Stahl und Eisen," 1941, 61, pp. 266–272.

at 250° C., a still, salt bath has a cooling effect equal to that of a still, oil bath at 20° C. This is due to the vapour films, formed in an oil-bath, interfering with rapid transference of heat from the hot steel to the liquid oil. Moreover, the use of oil at 250° C. is impracticable. The curves in Figs. 36 and 37 show, not only the relative cooling rates in various oils and in molten salt, but also the effect of raising the temperature (a) of the salt bath and (b) of the steel quenched therein.

Quenching speed is increased either by a motor-driven device, which circulates the molten salt, or by a pump which forces multi-jets of molten salt against the work held in fixtures.

A salt-bath quench is known to prevent quenching-cracks and minimize dimensional changes.

By using a salt-bath for each operation, viz., hardening, quenching and tempering, a considerable saving in time—of the order of 50%—may be effected.

It is not implied that the isothermal transformation method will supersede the conventional quenching plus tempering procedure under all conditions. It is a method which confers certain combinations of properties—on heat-treated steel—not obtainable by conventional procedure.

A product of equal hardness, but with greater toughness and ductility is obtained by the new method. As in the case of the older method, the composition of the steel, its prior treatment and the maximum thickness of the piece, influence the result. In cases where S-transformation curves are available for various steels, the information which they provide, indicates whether the application of the isothermal method will or will not be practicable.

The use of molten salt baths in this method of treatment provides the following advantages :—

(1) Uniformity of bath temperature, ensuring uniformity of austenite formation on heating or of austenite transformation on cooling.

(2) Low temperature gradient between temperature of bath and temperature at centre of piece being heated, ensuring uniformity of grain-size and of transformation-product.

(3) Minimum surface oxidation of steel pieces.

(4) Relative rapidity of cooling-rate—with bath temperatures above 200° C.

(5) Amenability of molten salt to agitation, circulation, or jet-projection by pumping through small orifices.

(6) Amenability of quenching bath to cooling by air-blast and agitation, whereby the heat introduced by the hot steel may be counteracted.

Molten lead has the advantage, over molten salts, of possessing higher thermal conductivity and therefore of promoting a more rapid rate of heating of steel articles. This advantage is offset in the case of articles of unequal section by the increasing risk of distortion with rapidity of heating. Another factor contributing to the comparatively slower rate of heating in salt baths is the sheath of frozen salt which encases the cold steel when it enters the bath. Before the heat can be transferred to the steel it must first remove this sheath by raising it back to melting point and supplying its latent heat

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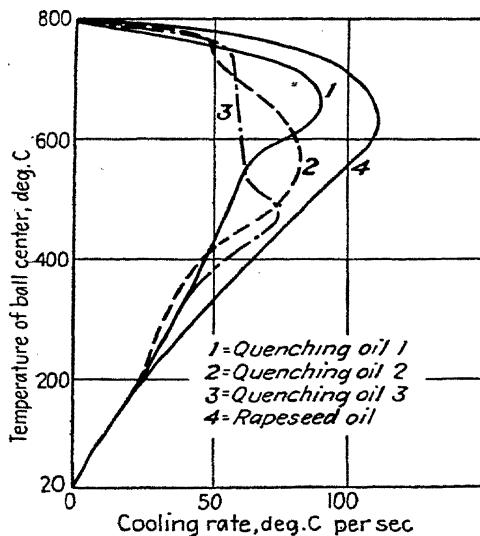
FIG. 36.   QUENCHING EFFECTS OF MINERAL OILS AND RAPE-SEED
OIL (AT 20° C.) ON A STEEL BALL (0·47-IN. DIAM.), INITIAL TEMPERA-
TURE OF WHICH WAS 800° C.
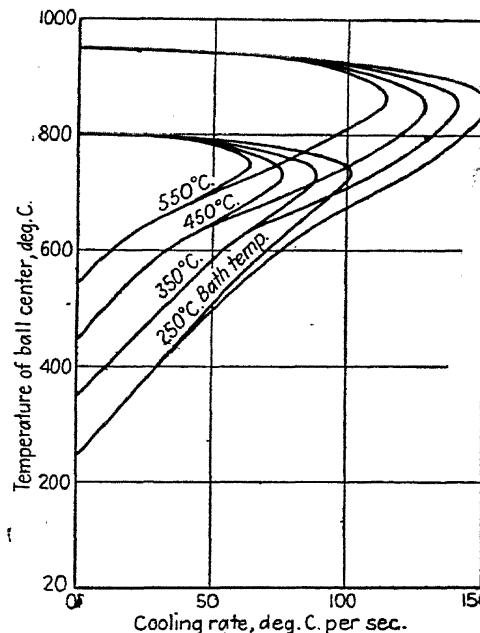
FIG. 37.   QUENCHING EFFECTS OF MOLTEN SALT (AT FOUR DIFFERENT
TEMPERATURES) ON A STEEL BALL (0·47-IN. DIAM.), INITIAL TEMPERA-
TURES OF WHICH WERE 800° AND 950° C.

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of crystallization. This takes time, allowing the steel to heat up more slowly. Owing to its low freezing point and low latent heat, lead does not cause such an appreciable lag in heat transference to the steel.

The chief application of the lead bath nowadays is in the heat treatment of tools or articles, part only of which is required to be fully hardened or separately tempered. The lead bath is convenient for such localized treatment.

Parts with teeth or thin projections should be pre-heated before putting into molten lead.

After heating, remove any adhering lead or dross with a stiff wire brush before quenching. Otherwise soft spots will occur.

Various coatings may be used to prevent lead adhering to steel tools. They are applied by brushing on or by dipping. The following coatings have been found to be effective :—

Strong solution of salt in water.

Soft soap.

Blacklead mixed with water.

Paste of blacklead mixed with flour.

Lamp black.

Mixture of lamp black and linseed oil.

Solution of 1-lb. of potassium cyanide in 1 gallon of boiling water (use cold).

Common whiting mixed with water.

A mixture of equal parts of fine corn meal and common salt. Tools are dipped in this after first dipping in water ; they are then heated until the coating " melts," then again dipped in the mixture, after which they may be heated for hardening.

After coating with any of the above, the tools should be allowed to dry before placing in the lead bath. Moist coatings will cause the molten lead to fly. Stirring occasionally is necessary to keep the molten bath uniform in temperature.

Rapidity of quenching is affected by adhering films of salt or lead or lead oxide. Salt films, being soluble, rapidly disappear in the quenching bath (if of water). Salt baths possess the following advantages over lead-baths in the hardening of files, viz. :—

(1) No " laying on " compound is required.

(2) Tangs need not be " drawn " (tempered) after hardening.

(3) Scouring is unnecessary, unless a high finish is wanted.

(1) One of the disadvantages of the lead bath is its ability to decarburize thin sections and edges. To prevent this, it is necessary to cover the teeth of the files with a " laying on " compound of carbonaceous matter. This coating is laid on wet and must be dried (*e.g.* in a blacksmith's hearth) before immersion in the lead bath.

Decarburization does not occur in a molten salt bath containing sodium cyanide and no " laying on " compound is required.

When file blanks are forged, the surface layers become decarburized and coated with scale. The scale is removed by grinding but some of the decarburized skin will remain. The " laying on " compound will recarburize such areas, unless it falls off during handling. In that event soft patches

may be left on the hardened file. This cannot happen in the cyanide bath, which effectively recarburizes any areas deficient in carbon when the file enters the bath.

(2) In order to keep a file vertically immersed in a lead bath, some mechanical device must be used to hold the file (of lower density than the lead) below the surface of the bath. So deeply must the file be immersed that the tang becomes heated by contact with the red-hot layer of coke which covers the lead in order to prevent the formation and adherence of dross on the tools. So, when the file is hardened, the tang is hardened too, and must be softened by tempering (a separate operation).

In a cyanide bath the files, by reason of their higher density, are readily suspended or held in a vertical position with the tang clearly projecting above the surface. Consequently the tang is never heated to the critical point, does not harden and therefore does not require a separate tempering operation.

(3) Scouring is necessary in order to remove adhering " laying on " compound from between the teeth of the files heated in molten lead. Files hardened from salt-baths are clean after quenching and do not need scouring. Fine cut files are sometimes mottled but have a brilliant silver-grey finish after a very light scouring operation, as compared with the dull grey appearance of files which have undergone the heavy scouring necessitated by the lead-bath hardening.
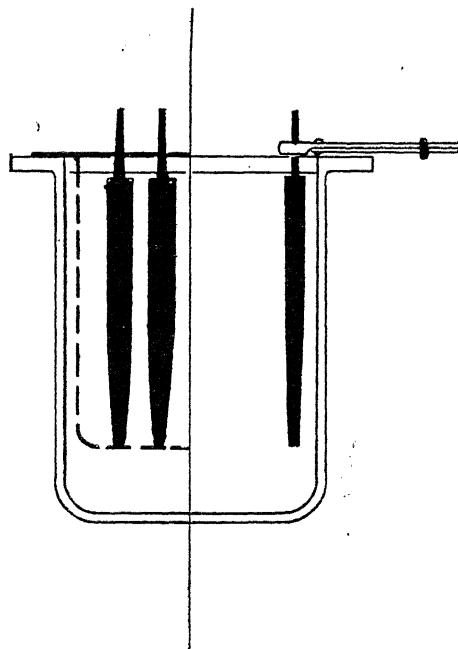
FIG. 38. METHODS OF SUSPENDING STEEL FILES IN A MOLTEN LEAD BATH.

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The slower rate of heating in salt-baths may be compensated by increasing the number of files in the bath at one time, say nine (instead of six in the lead bath).

Two methods of holding the files in the bath are shown in Fig. 38. In the left half of the illustration, a perforated basket is shown, of such a length that the tangs of the files remain outside the bath. In the right half of the illustration is shown the method of suspending a file by gripping the tang in a pair of tongs resting on the flange of the pot. Files thus suspended may be arranged all round the edge of the pot.

Owing to the high density of lead, extra strong pots are required. These are of cast iron, about $2\frac{1}{2}$ in. thick. Molten salts have only one-seventh the weight of lead and pots of $\frac{1}{2}$ in. thickness suffice to hold them. They may be made from lengths of mild steel pipe (solid drawn) the bottom being formed by welding on a piece of steel plate. A flange, 3 in. wide, should also be welded round the top. The welding must be done carefully. The cost of such a pot is considerably less than that of the heavy cast-iron pots used for lead.

Soft Spots.

Heat should be abstracted from a given article, not only at a sufficiently rapid rate, but at a uniformly rapid rate throughout the mass. If certain parts of the mass are prevented from cooling fast enough they will not harden and their location will be indicated by soft spots. The quenching of too many articles at a time often results in a heap lying on the bottom of the quenching tank, raising the temperature of the liquid in their neighbourhood to an extent which reduces its quenching severity and causes irregularity in its action. Steam collects in blind holes or in pockets, in articles of intricate shape, and is responsible for a local slowing-down of the cooling rate, resulting in soft spots. Continuous agitation during quenching will obviate this trouble.

Oils of high viscosity, as already indicated, can be responsible for local retardation in cooling rate. Oils containing fatty matter may undergo local saponification and thickening, by reaction with salt adhering to steel which has been quenched from a salt bath. Local thickening can also occur with oils of high volatility as already explained. Adhering scale, on articles heated in an oxidizing atmosphere, may affect the cooling rate and cause soft spots.

The finer the grain-size of a steel, the greater its tendency to suffer from soft spots.

Detection of Soft Spots.

Soft spots are revealed as dark-etching areas, when the steel, after polishing and degreasing, is etched in a solution of 10 % nitric acid in alcohol.

Tempering.

Sometimes called " drawing," this operation is the reheating of a hardened steel to some temperature below Ac_1 followed by any desired rate of cooling.

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A uniformly slow rate of heating, with accurate temperature control is necessary for tempering operations. After holding for the required length of time at the temperature for tempering, the articles are removed and allowed to cool in still air, or, in some cases, quenched in oil or water. Straight carbon steels require only 15 minutes or so at the tempering temperature, the time varying with dimensions and quality, but alloy steels require much longer periods if the maximum impact values, for a specified tensile strength are to be obtained. Such steels also often require higher tempering temperatures. As the temper colours are different in different alloy steels, it is better to use an oil bath or, for higher temperatures, a molten salt bath* pyrometrically controlled. For those inexperienced in judging temperatures by surface colouration effects, the tempering bath or the forced-air-circulation furnace, temperature-controlled, is safer for all steels. The rate of heating must be slow, for reasons already given in discussing hardening.

Although any mistakes in heat-treatment can be rectified by re-normalizing and again hardening and tempering, some specifications put a limit to the number of times this may be done. Correct treatment can be established by careful trials, providing the composition of the steel does not subsequently vary.

From what has already been discussed in connection with the residual stresses in hardened steel, it will be understood that the sooner these stresses can be relieved the better, as, otherwise, the steel may not escape their destructive action. The tempering operation has, as one of its objects, the relief of internal stresses, therefore no time should be lost, after hardening, in subjecting articles to the tempering treatment. Articles of intricate shape should even be transferred—to the tempering bath or furnace—whilst still warm to the touch. Allowing a steel, after hardening, to remain untempered for a period of time, at room temperature, often results in cracking. Transference, while still too warm, may also cause cracking.

In the case of articles made from sheet or strips (e.g. saws), warping may be prevented by placing the articles between heated dies, under pressure. This method of tempering has three objects, viz. :—

(a) stress-relief.
(b) removal of brittleness.
(c) flattening.

A gas-heated or electrically-heated plate may be used, if tempering baths are not available. After cleaning and brightening the surface of the tool or component, it is placed on a hot plate and allowed to remain until the required temper colour film appears (see p. 70), being then removed and cooled (usually in water). Small tools may be tempered in hot sand. Articles of complicated design should not be tempered by either of these methods, owing to the irregularity in results. Chisels, snaps, setts, etc., may first have their shanks heated and quenched in oil. The cutting end for a distance of $\frac{1}{2}$ in. to $1\frac{1}{2}$ in. is then heated to the hardening temperature and quenched as described above. Some of the heat from the cutting end will be conducted to the previously hardened shank and will temper it. This oil-quenching and tempering treatment of the shank increases its durability, rendering it less liable to spall or split under the severe hammering it receives in use.

* See page 71.

Tools of intricate shape should not be put straight into a tempering bath, but should be first pre-heated. Alternatively they may, like large articles, be placed in the bath while it is at a comparatively low temperature. The bath may then be raised to the temperature specified for the tempering of the particular tools under treatment.

Hardened tools may " clink " in the centre if not brought gradually to the tempering temperature.

" Clink " may be defined as the condition of a steel when it is internally weakened by severe internal stresses or even incipient fracture. This condition is not outwardly visible, but renders the steel liable to break under abnormally low stress, the fractured surfaces having a characteristic " cup-and-egg " appearance.

Tools should not rest on the bottom of the container holding the tempering medium. By means of a false bottom or by some method of suspension, e.g. a wire basket, a clear space of not less than 1 in. should be maintained between the work and the bottom of the bath. An oil bath is strongly recommended for the tempering of dies. Very large articles may benefit by cooling with the bath, thus minimizing risks of distortion.

Lead-tin alloys give a lower range of tempering temperatures than pure lead. Pure lead freezes at 327° C. Pure tin freezes at 232° C. An alloy of 61·86% tin, 38·14% lead freezes at 183·3° C. (eutectic alloy). Temperatures at which freezing commences vary fairly regularly between 327° C. and 183·3° C. on the lead side of the eutectic and between 232° C. and 183·3° C. on the tin side.

TABLE VIIIA. *

TEMPERING TEMPERATURES, ASSOCIATED TEMPERING COLOURS AND CHIEF USES.

Tint of oxide on surface of steel.	Centigrade.	Fahrenheit.	Suitable for.
Dark blue	316°	600°	Hand saws.
Blue	293°	560°	Fine saw blades, augers, boilermakers' snaps, chisels, smiths' tools and cold setts.
Bright blue	288°	550°	Watch springs, swords.
Purple	277°	530°	Table-knives, large shears and wood-turning tools.
Brown beginning to show purple	266°	510°	Axes, planes and wood-working tools.
Brown	254°	490°	Scissors, shears, cold chisels, large drills, shear blades, punches and wood-cutting tools.
Golden yellow	243°	470°	Penknives, hammers, taps, reamers, large lathe, planing and slotting tools, small drills, screwing, stamping, and cutting dies and miners' drills.
Straw	230°	446°	Razor blades.
Pale yellow	221°	430°	Small edge-tools, small lathe, planing and slotting tools.

N.B.—The above data apply to straight carbon steels only. They are not true for alloy steels and are, in any case, affected by variation in mass and time of heating.

* *Edgar Allen News, July, 1941.*

Baths for Tempering.

The advantages possessed by salt baths in hardening operations also apply in the case of those used for tempering. The useful range of temperature for tempering baths must be from 200° C. to 650° C. For lead baths the working range is from 350° C. to 800° C. Tempering below 200° C. may most conveniently be done in oil as salt baths melting below 200° C. are costly. Plain carbon steels are tempered above 200° C. in molten baths of mixed nitrates. A temperature of 550° C. should not be exceeded, as, above this temperature, the steel surface is liable to become scaled or pitted, by the prolonged action of molten nitrates. Steel which has been hardened from a molten cyanide bath should be washed and dried before immersion in a nitrate bath to avoid risk of explosion.

As a general rule, the tempering of nickel-chromium steels may be done in the bath which is used for heating for hardening.

Prevention of Rusting after Hardening.

The sheath of salt which encases a steel article when removed from a molten salt bath, is readily removed by water-quenching. The salt is not only removed but is dissolved by the water. This is not so when oil is used for quenching, a considerable quantity of salt being retained on the steel. Unless washed in running water, the salt, which is hygroscopic, will encourage rusting, by absorbing and retaining atmospheric moisture. Washing must be very thoroughly done, so as to remove the least trace of adhering salt from the steel. Salt may easily be retained in articles having blind holes or of so complicated a shape as to provide pockets, in which salt remains. All such retained salt must be completely removed by washing in boiling water, followed by a thorough washing in running cold water. This washing also removes from the steel, practically all adhering grease or oil which otherwise might have protected it from the atmosphere. Immersion in boiling water containing 10 to 20% soluble oil, followed by drying, will afford some degree of protection. A more lasting effect is obtained for parts requiring to be stored for a long time or despatched on long journeys, by means of anti-rust preparations, such as lanoline solutions.

Modern Equipment.

The temperature of the load of steel in the furnace should be determined by inserting a thermo-couple (asbestos covered) into a piece of rolled steel in which a hole has been drilled to take the couple. This is then put as nearly as possible in the centre of the load. Thus the temperature at the centre of the load is taken and the risk of false indications such as would have been obtained by having the couple outside or nearly outside the load, is avoided. Moreover, the couple is kept out of contact with the packing materials, the reducing nature of which might vitiate the reliability of the couple. The couple should be connected to a recorder, which will give a continuous record of the temperature of the load. A couple connected to an indicator may be used to provide a check on the temperature at other points in the furnace. Agreement between the readings of the two couples will show uniformity. Any disparity will direct attention to non-uniformity which must be rectified.

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The following is a brief description of one of the City cf Birmingham Gas Department's new Heat Treatment Sections, of the Industrial Research Laboratories.

The lay-out of the plant is based on many years' experience in the heat-treatment of materials, and it incorporates the essentials.

The building is of brick with a patent glazed roof, which, of course, gives a well-lighted shop. The floor is laid with a special granolithic material and occupies approximately 12,000 sq. ft. Offices, stores, etc., cover about 70 feet of the 130 feet frontage, the remainder being taken up by a large Receiving and Despatching Bay.

A battery of gas-fired furnaces is installed in a continuous line along one side of the building.

These are varied in type and size, from large recuperative oven furnaces used for carburizing, annealing, and general heat-treatments of drop forgings, tubes, etc., to the smaller types of natural-draught furnaces for the hardening of beds, dies, etc., and furnaces for the hardening of high speed steels, as well as salt-bath furnaces for high and low temperatures.

The whole of the shop is ringed by gas supply pipes, 6 inches diam. on two sides and 4 inches on the other two. Each side of the building can be isolated by valves. This allows for additional apparatus to be fixed or repairs to be undertaken without interfering too much with the general working of the plant.

A 4 inch air supply pipe runs parallel to the 6 inch gas main to supply air for cooling purposes, such as for the air hardening of steels, etc. The air is supplied by two belt-driven blowers of 10,000 cubic feet per hour capacity at 2 lbs. pressure each, to be run singly, one being a standby in case of breakdown.

There are two large water tanks, one of 3,000 and the other of 1,200 gallons capacity, used for the quenching of the material being heat-treated. These tanks are supplied from a 7,000-gallon storage tank on the roof of the building, a 4 inch supply pipe being taken from the tank to discharge above and into the quenching tanks, controlled by valves.

The overflow from the tanks is arranged by means of two 4 inch pipes, set near the top of the tank and a channel in the floor of the shop 2 feet wide by 1 foot deep, to a reservoir under the shop floor, holding 11,000 gallons of water. From here it is pumped back to the tank on the roof by a pump of 220 gallons per minute capacity, another pump of the same capacity acting as standby. Provision is made to fill the storage tank on the roof from the town's main, a 2½ inch copper supply pipe, fitted with a gate valve, enabling the tank to be filled at will. The underground reservoir is connected direct to the main sewer, both as regards overflow and emptying for cleaning purposes.

Two large oil tanks of 1,750 and 1,000 gallons capacity, again for quenching, are connected to oil coolers by means of 4 inch pipes.

Each cooler has its own motor and pump ; one cooler uses water as the cooling agent, the oil being pumped through aluminium bronze tubes, the water for cooling purposes flowing round the tubes.

The other cooler uses air as the cooling agent, the air being supplied from a fan at the top of a tower, and passing down through a duct to the bottom of the cooler. Oil is pumped to the top of the tower and then passes down through several grids of copper foil, then out through a 4 inch pipe to its assigned tank.

There are also two small quenching tanks, one for water, the other for oil, each holding 200 gallons. Both tanks are connected to a refrigerator. The low temperature of the quenching media in these tanks allows for rapid abstraction of heat and is applied to the quenching of material required to have a special hardness.

All the quenching tanks are situated near to the furnaces, about 3 feet to 6 feet from the front of the smaller furnaces, and 10 feet to 12 feet from that of the larger furnaces. This is, of course, necessary, so that work being quenched can be transferred from furnace to quenching media with as little delay as possible.

For removing work from the receiving bay, and for loading and unloading furnaces, there is a 1-ton electric hoist conveyor which travels completely around the shop, and also loops over all the quenching tanks.

Several sets of loose bogies for holding blocks and quenching baskets are also used over the quenching tanks, in addition to the electric hoist.

All furnaces are fitted with thermo-couples which are connected to temperature indicators or recorders by means of compensating leads. Automatic control enables certain furnaces to be raised to, and maintained at, any predetermined temperature, the gas supply and waste gas flue being opened or closed by the controller as the furnace temperature rises above, or falls below, the desired temperature.

The many types of steel in use to-day, and the many specifications to be met, call for close co-operation with the Physical and Mechanical Testing Section and also the Chemical and Metallurgical Section of the Industrial Research Laboratories which are provided with the testing machines and other necessary equipment for facilitating the fulfilment of specification requirements and for the investigation of all problems arising during treatment.

Machining.

Machining is a form of cold-working and internal stresses occur in machined articles to some extent. Internal stresses are produced before hardening, in various ways—in rolling mill or press-shop—and it is essential to normalize before the final hardening operation. This treatment should enable hardening to be done without distortion.

Wherever it is possible to machine after heat-treatment, a great advantage is gained, owing to the elimination of the distortion which heat-treatment often creates. Thus rolled bars may be heat-treated to possess the required mechanical properties, being subsequently cut and machined to final size. The upper limit of tensile strength to which it is practicable to machine heat-treated steel, is approximately 66 tons/in^2 although a higher limit is technically possible. Where the required strength and hardness of the finished articles render machining economically prohibitory, it is desirable to adopt the following procedure, viz. :—

(1) anneal,
(2) machine roughly to size,
(3) heat-treat,
(4) finish-machine to final dimensions.

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Articles of complicated form are so prone to undergo distortion, when heat-treated, that some degree of finish-machining is imperative. High alloy steel forgings are usually too hard to machine until they have been annealed.

Articles which have been machined (*e.g.* drilled, planed, slotted, etc.) are in a state of internal stress in the immediate neighbourhood of the machined surface. They are rendered less susceptible to cracking in the hardening operation if they are first normalized.

When an article, such as a steel shaft, distorts, as the result of heat-treatment, it is often cold-straightened. Subsequent machining, such as the cutting of a keyway, upsets the balance of internal stresses and distortion again occurs.

When it is desired to remove the decarburized skin from round bars by machining, correct centring is necessary to ensure removal of metal to an overall uniform depth of cut. The usual machining allowances are as under :—

1 inch diameter and under	$\frac{1}{32}$ inch
1 ,, ,, to 2 inch diameter	$\frac{1}{16}$,,
2 ,, ,, ,, 3	
3 ,, ,, ,, $4\frac{1}{2}$	
,, ,, ,, 6	

Grinding.

All general cutting-tools should be machined or ground before hardening, in order to remove the decarburized skin which is always present on steels after being hot-forged or hot-rolled. This skin is incapable of hardening and may even interfere with the proper hardening of the steel underneath.

The heat generated by dry grinding can have marked influence on the structure of hardened steels, therefore wet grinding—in which operation also tools are cooler to handle—is to be preferred.

Grinding pressure, if excessive, or too suddenly applied, can be the cause of surface cracking.

Choice of the correct type of grinding-wheel is important. Wheels of coarse grit and soft bond minimise risk of cracking when grinding hardened steels. The heat generated by friction raises the surface temperature of a tool and the highly-localized expansion inseparable from this rise in temperature must produce internal stresses, often great enough to produce cracks. Even annealed high carbon-steels may be surface-hardened by grinding as a result of (*a*) rise in surface temperature due to friction ; (*b*) rapid abstraction of heat from the surface layer by the cold mass of metal beneath and/or by the quenching action of the coolant. Hard spots, thus formed, cause trouble in machining and are often associated with surface cracks.

Surface *softening* of a hardened steel can also occur as the result of excessive grinding. A Rockwell C. hardness of 67 can be reduced to 55 in this way. The heat generated by friction has a localized tempering effect.

It is sometimes thought that great hardness of a steel, whether due to composition or to heat-treatment, may cause increased difficulty in grinding. This, however, is not the case.

PRACTICAL EXAMPLES OF HEAT TREATMENT TECHNIQUE

MILL-HARDENING

J. Winning* draws attention to the fact that, in some German steel works, a fully developed system of direct heat-treatment from the rolling mill or forge has come into use. In such a system, results show that, in addition to the saving in time and the saving in fuel by utilizing the heat still in the metal as it leaves the rolls, superior physical properties are obtained in the product.

Care is taken that the steel is still hot enough to harden at the finishing pass.

Uniformity of temperature in the bars or sections is obtained by immediate transfer to a soaking-furnace, in which equalization is obtained in about half an hour. Such a furnace is economical to run, owing to the heat-input derived from the hot bars.

Care must also be taken that the rolled sections are not allowed to reach grain-growth temperature, with the accompanying risk of embrittlement by grain-coarsening.

Long sections are cut into suitable lengths at the hot saw, passed to the soaking-furnace on mechanical skids, with very little loss of heat. They are quenched in water or oil direct from the furnace, tempering being carried out in a separate furnace on continuous production lines. For bars and sections, required to be long, a tunnel type of soaking-furnace is used, with a pusher or walking-beam mechanism, which automatically delivers them into the quenching tank. Owing to machining difficulties presented by bars in the heat-treated condition, the process is likely to find its greatest application in the case of steels in the 50 to 60 tons./sq. in. range, which do not present impracticable machining troubles.

SOME POSSIBILITIES OF HEAT-TREATED LOW-CARBON STEELS

Merle Rendell † states that, by drastic quenching, low carbon steel (0·20% C.) can acquire yield points as high as 58 tons/sq. in., and tensile strengths as high as 84 tons/sq. in.

The process, invented by G. F. Nelson of Berkeley, California, U.S.A., consists in quenching from a temperature of approximately 925° C. in a quenching medium of sodium or potassium hydroxide solution (35%) at room temperature. The quenched steel can be cold-rolled and possesses many of the desirable qualities of alloy steels, also of high-carbon steels, without some of the latter's objectionable qualities.

HEAT-TREATMENT BEFORE MACHINING

W. S. Craig‡ has described the metallurgical practice in the manufacture of rifle barrels in a small arms factory.

The continuous heat-treating section is concerned primarily with the conditioning of steel to obtain consistent machineability and necessary physical properties in parts which cannot be heat-heated after machining.

* " Mech. World." September 3, 1943, pp. 259–260.
† " Heat Treating and Forging." October, 1943.
‡ " Canadian Metals and Metallurgical Ind." December, 1943, pp. 20–23.

The rifle barrel is a typical example ; since very close bore tolerances must be held, treatment following drilling, reaming, rifling and lapping operations is impossible, so the barrels are hardened and tempered in the forged state to ensure uniform hardness and required physical properties.

An interesting feature of the barrel-hardening equipment is the use of a hydraulic quenching machine which, by means of rotating contour rollers, straightens the hot barrels and, while spinning, quenches them, the result being straight, uniformly hard forgings which can be held to close tolerances, thus saving material and machine tool capacity. Continuous furnaces used are electric radiant-ring units with chain or link-conveyors, having capacity to harden approximately 2,500 lbs. per hour per unit. Quench tanks, with capacities from 500 to 3,700 gals., have forced circulation, temperature control and conveyor discharge where necessary. Central oil systems serve all units to ensure adequate quenching.

MODERN METAL CUTTING

C. J. Oxford* has drawn attention to the importance of the following factors in metal cutting, viz. :—selection of the correct tool for the job, rigidity of machines, fixtures and the work piece itself, tool maintenance, re-sharpening, selection of proper grinding wheels for tool reconditioning, handling of tools, storage and transportation and finally salvage of worn tools and tool-materials.

With the wide variety of metals to be machined, with varying heat-treatments of these materials and the varying nature of the cuts to be made, there cannot be one standard type of tool that will perform to best advantage on every job. For example, twist drills are required to operate efficiently on all types of material from the free-cutting magnesium alloys to the toughest and hardest armour plate. Extremely deep holes may have to be drilled as well as the medium and extremely shallow ones. To meet all situations several types of twist-drill have been developed, which may be considered semi-standard. In addition to the regular type, other types are described.

As regards handling, attention is drawn to the delicate nature of tools, the destruction of which may easily ensue as the result of careless handling. Hardened tools are usually somewhat brittle, so that they should not be expected to be immune to such practices as dropping on to hard floors or hammering into tool-holders where there is a tight fit.

HEAT-TREATMENT AND TOOL DESIGN

A. S. Eves suggests that of all the tools designed and started through the tool-room, not more than 25% ever succeed in finally reaching the job and there achieving maximum productivity and life. In the selection of tool steel there is no excuse for specifying case-hardening, when the steel contains much carbon, or for using water-hardening steel for delicate tools when there are so many varieties of oil-hardening tool steel, or for choosing an alloy steel to give a higher degree of hardness than was intended for it. Nor should the heat-treater be expected to produce springs from steel stamped out of cold-rolled strip. In a commercial heat-treating plant (the one in question handles 2 tons of high speed steel and 20 tons of other steels each day) it is absolutely essential to know the composition of every tool if the best possible results are to be obtained. The spark-test is not a safe guide, as

certain oil-hardening steels give a spark like that of a straight carbon steel. It is suggested, in order to conserve man-hours and avoid making scrap, that the following points receive careful attention, viz. :—

(1) Tools and components should be designed not only to do the job, but also to undergo successfully, the hardening treatment specified.
(2) Steels should be selected that are not only correct for the intended service, but also satisfactory as regards physical properties.
(3) The right treatment must be chosen, bearing in mind the difference between heat-treating for hardness and for toughness.
(4) At least $\frac{1}{16}$ inch should be removed from all sides, by machining. Strains caused by hogging should be avoided.
(5) The hardener should be fully and correctly informed as to the kind of steel supplied. In case-hardening he should know the depth of case required and whether the work is to be hardened or left soft.
(6) The hardener should be informed as to the intended use of the tool or component.
(7) Test pieces should be supplied similar in size and shape and composition to the work which they represent.
(8) Forgings should be first rough-machined, then normalized *and* annealed, before machining to size.
(9) Care should be taken to ascertain whether the hardener possesses the degree of skill and the kind of experience requisite for securing the best results.
(10) Care should be taken to ensure the hardener being provided with sufficient modern, suitable equipment.
(11) No new process which possesses greater advantages, should be overlooked.
(12) The pitfalls associated with straightening, blasting, tumbling and pickling tools of delicate shape and design should be avoided. For example, straightening may be obviated by selective hardening.
(13) Finishing-grinding must be carefully done, so as to avoid grinding-cracks.
(14) Correct rakes must always be specified.
(15) Grinding must be done in good time.
(16) Correct speeds and feeds must be chosen for the heat-treated tools.
(17) The condition of the steel as received must be considered. It will pay to normalize or anneal rather than waste tools.
(18) Tools which work-harden may be improved by re-tempering.
(19) Reclamation of tools after use, should not be overlooked.

Flame Hardening.*

This method was first used and adopted by the railways in 1927 for rail-end hardening to reduce rail wear produced by the constant battering of the rolling stock travelling over them. The operator carries his oxy-acetylene equipment on a suitable rail hand-truck and travels along the line, inspecting all joints. The rail-ends which are worn, are built up by welding and then flame-hardened. New rail-ends are also hardened. The operator heats the rail surface at the joint to a temperature above the critical point for a distance of 3 inches from the end. The heated surface is then hardened by water-quenching and tempered to a lower degree of hardness (350 Brinell) by

* *Further information on this and other surface-hardening processes will be found in Vol. III.*

reheating. The correct tempering temperature is determined by pyrometer or by the use of a lead-tin alloy whose melting-point corresponds to the temperature required.

For areas of extended length this method of unit-area hardening is not practicable. Therefore, progressive hardening is employed, whereby multiple heating flames are moved progressively over the surface under treatment, followed by a suitable stream of water for quenching or compressed air for air-hardening steels. By using different quenching media, different degrees of surface-hardening were obtained but not differing greatly from the maximum hardness obtainable on the steel under treatment.

Stephen Smith* has recently developed a " simultaneous " flame-hardening and tempering process. Unlike furnace hardening, wherein three separate operations are involved (heating, quenching and tempering), this process combines three into one.

The " simultaneous " process embodies the use of a standard multi-flame oxy-acetylene flame-hardening tip with a quenching jet incorporated, following the flames. A soft multi-flame heating tip is employed for the tempering operation at a definite distance in the rear of the quenching jet. This tip is attached to the flame-hardening tip by a suitable fixture, thus making a unit assembly. The unit is moved progressively along the surface to be treated at a definite speed for a desired depth of hardness. By controlling temperature, pressure and volume of the stream of water the severity of the quench may be varied, as also the intensity of the tempering flame (by varying the gas pressure). A baffle plate behind the quenching jet prevents excessive splashing on to the tempering flame, thus avoiding erratic tempering effects. Uniform hardening temperature is ensured by the unvarying temperature of the oxy-acetylene flame. With all phases of the process under positive control, uniform results are assured and warrant large-scale production of identically hardened parts. A number of such production installations are now in operation.

It is suggested that the most fertile field of application of the process will be to confer definite surface hardness on parts subjected to heavy impact and fatigue stresses, such as axles, torque shafts, propeller shafts and oil-well drill stems.

Classification of Tool Steels.

An American classification of tool-steels divides them into :—

(1) **Hard Steels.**

(a) Those capable of being hardened to over Rockwell 61 C.

(b) Those containing over 0·55% carbon, even if their hardness is less than 61 C. (Rockwell).

(2) **Tough Steels.**

Those containing less than 0·55% carbon, providing their hardness does not exceed 61 C. (Rockwell).

Impact Tools.

It is imperative that tools such as machinists' hammers or steel number or letter punches should be correctly hardened. Shallow or imperfect

* Stephen Smith, Canadian Metals and Met. Ind., March, 1943.

hardening will cause hammers to spall or split. Flying fragments from such tools when in use are highly dangerous.

The heads of punches will mushroom, if too soft. Bevelling will lessen this trouble and the danger of flying fragments. If the body of the punch is too soft it will bend in use. If and when such a tool breaks, the broken parts will be projected with greater force than in the case of the breakage of a harder and more brittle punch. All punches softer in the body than 44 C. Rockwell, will bend in use and may easily fly out of the operator's hand when struck. The marking end of a punch will have a longer life when correctly hardened.

The Sheffield (England) classification of steels in six degrees of hardness (or " temper ") is given in the following list taken from the *Edgar Allen News* of July, 1941.

The temper-numbers indicate only the relative hardness of the steel ; its quality is denoted by a quality or class-letter alone.

Temper No. 1 (1·35% carbon). Suitable for extra-hard planing-, slotting- and turning-tools, drills, etc.

Temper No. 2 (1·20% carbon). Suitable for lathe-tools, drills, and small cutters.

Temper No. 3 (1·05% carbon). Suitable for large turning-tools, cutters, taps, reamers, drills, punches, blanking tools, etc.

Temper No. 4 (0·90% carbon). Suitable for cold chisels, blades for hot shearing, hot sates, taps, special miners' drills, etc.

Temper No. 5 (0·75% carbon). Suitable for chisels, sates, black-smiths' tools, blades for cold shearing, etc.

Temper No. 6 (0·60% carbon). Suitable for boilermakers' tools, hammers, miners' tools, etc.

Chapter 3
HARDENABILITY

It has been pointed out that, in order to obtain optimum results as regards mechanical properties, by heat-treating a steel, it should be *fully* hardened before the final tempering operation.

Full hardening implies obtaining maximum hardness in the whole piece, but (except in very small sections) with straight carbon steels this is impossible, because the rate of cooling at the centre must always be slower than at the surface. The bigger the section the greater this disparity, and even water-quenching of the most drastic kind will leave the centre of the steel less hard than the maximum. In pieces of complex shape, water-quenching, owing to risks of distortion, or even cracking, may not be permissible.

A milder form of quenching, *e.g.* oil-quenching, must be adopted in such cases, with results still further removed from the ideal full hardening. Results obtained by Grossmann* illustrate these points (Table IX). The steel used for these experiments had the following composition, viz. :—

Carbon	Manganese	Phosphorus	Sulphur	Silicon
0·48%	0·60%	0·022%	0·016%	0·17%

TABLE IX.

Diameter of Piece.	Hardness at Surface of Piece Rockwell Hardness No. (C – Scale).	
	Water	*Oil*
0·5 in.	59	29
1·0 in.	58	29
2·0 in.	41	28
3·0 in.	35	23
4·0 in.	30	23
5·0 in.	24	18

The hardness at the centre of each piece was much lower than the surface-hardness values given above. Even in the case of the water-quenched $\frac{1}{2}$ in. diameter specimen, the hardness number at the centre was only 51 as compared with 59 at the surface.

In certain applications, it is not necessary for the hardening effect to penetrate right through to the centre ; steels may indeed be required to have a high surface hardness and a softer centre. In shafts and members subjected to torsional stress only, such a condition gives better resistance to torsional fatigue.

Quite a number of automobile steels give reliable performance even though not fully hardened throughout.

* *Principles of Heat-Treatment.*

In those applications where such latitude is not permissible, *i.e.* where the fullest possible hardening penetration is required in small and large pieces or in pieces which must of necessity be less severely quenched, straight carbon steels must be replaced by alloy steels of such composition as will possess greater hardenability (susceptibility to hardening by quenching).

R. W. Roush* states that a great deal of emphasis is placed upon the hardenability test to-day, as a means of evaluating a steel. It is a rapid and convenient means of determining the hardening properties and a first indication of the possibilities of a steel, but to draw all of our conclusions on the basis of hardenability is to neglect the very important consideration of ductility. It is generally conceded that a steel will perform best if it has been fully hardened all the way through, to its maximum hardness, and then tempered back to the hardness desired, and we cannot hope to get the most out of a steel unless it is so hardened. It is therefore important to know *how* it will harden.

As tensile strength is closely proportional to hardness, it can be said that hardenability is a measure of strength, or rather of the capacity for strength, especially the strength of different portions across a cross-section. The elastic limit and endurance limit are proportional to hardness provided that initial hardening was complete, but are not always consistent after incomplete hardening. Elongation, reduction in area, and impact properties show a definite relationship to hardness and strength in the lower hardness ranges, but in the as-quenched condition and in the high-hardness ranges,
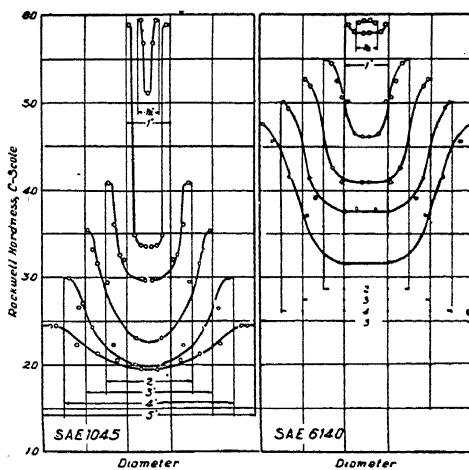
FIG. 39   FIG. 40

HARDNESS DISTRIBUTION IN VARIOUS SIZED STEEL BARS, WATER-QUENCHED.

FIG. 39. CARBON 0·48%, MANGANESE 0·60%.
FIG. 40. CARBON 0·42%, MANGANESE 0·73%,
CHROMIUM 0·94%, VANADIUM 0·17%.
(*Grossmann.*)

this relationship does not always hold, and there are many parts, such as gears, which are used in these conditions.

In such cases, laboratory or performance tests on actual parts are the only safe and sure means of proving the value of a steel. Although the hardenability test has proved its great value in framing National Emergency steel specifications and was probably the only one suitable, for the reasons given above, the real value of a steel should be confirmed and verified by laboratory and performance tests.

In Fig. 39 is shown a " hardness penetration diagram." Such a diagram is obtained by plotting hardness values—obtained on a ground and polished flat face of a circular cross-section of a hardened steel rod—against distances between edge and centre. This diagram clearly shows (A) diminution of hardness with increase in diameter of rod ; (B) disparity in hardness values between edge and centre.

By comparing the results of oil-hardening (Fig. 41) with the above results of water-hardening, it will be seen that similar effects are obtained, but with a far lower intensity ; in fact, true hardening has not occurred, even at the surface of the smallest ($\frac{1}{2}$ in. diameter) rods.

To lessen the falling-off in hardness from surface to centre and to raise hardness values all round, alloy steels are used.

In Fig. 40 is shown the hardness-penetration diagram for a water-hardened steel of the following composition, viz. :—

Carbon	Manganese	Phosphorus	Sulphur	Silicon	Chromium	Vanadium
0·42%	0·73%	0·027%	0·023%	0·25%	0·94%	0·17%

FIG. 41.   FIG. 42.
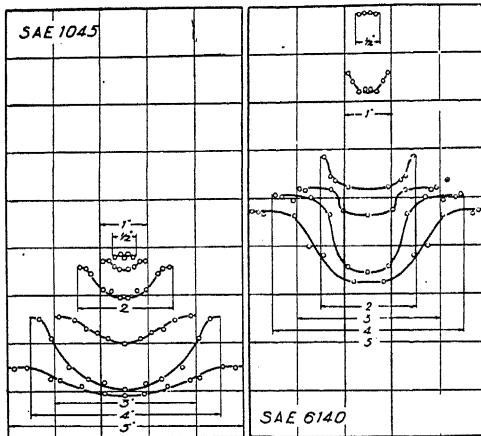
HARDNESS DISTRIBUTION IN VARIOUS-SIZED STEEL BARS.   OIL-QUENCHED.

FIG. 41.   CARBON 0·48%, MANGANESE 0·60%.

FIG. 42.   CARBON 0·42%, MANGANESE 0·73%, CHROMIUM 0·94%, VANADIUM 0·17%.

(*Grossmann.*)

The addition of chromium and vanadium have conferred greater hardenability. This steel will give full hardening effects in ½-in. diameter bars even when oil-quenched (Fig. 42) and higher hardness values for any given size, position and rate of cooling (the rate of cooling in oil is only one-third as fast as the rate of cooling in water).

When a hardened steel bar is fractured, two types of fracture may be seen, viz., a core, characteristic of tough steel, and an outer rim, characteristic of hard, brittle steel. There is a sharp line of demarcation between them. When ground smooth, polished and etched, the two zones are sharply differentiated, the outer zone being bright and the inner zone dark.
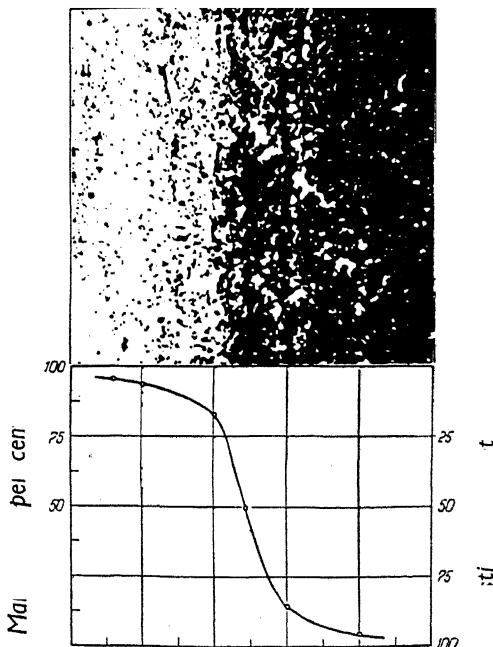
FIG. 43. PHOTOMICROGRAPH AND CHART SHOWING RAPID TRANSITION FROM PREDOMINANTLY MARTENSITIC (LIGHT) TO PREDOMINANTLY TROOSTITIC (DARK) STRUCTURE. (*Grossmann.*)

Hardness tests also reveal a relatively sharp change at the position of the change in type of fracture and in etching behaviour. In Fig. 43 is shown the cause of this change. The microstructure of the hardened zone is predominantly martensitic (light etching), that of the unhardened core predominantly troostitic (dark etching nodular troostite).

The abrupt change in fracture and in etching behaviour is therefore due to a sudden transition from martensite to nodular troostite. The *rapid* increase in the amount of nodular troostite occurs where there is about 50% troostite in the microstructure.

As it is easy to measure the 50% troostite amount, and as the greater the hardenability of a steel the longer will be the time taken by it during

quenching to form troostite, so the time required under standard quenching conditions, to form 50% troostite, becomes a convenient measure of hardenability.

It can be shown that the sudden fall in hardness (*see* Fig. 44) corresponds to a particular cooling time, and this cooling time, wherever experienced, will lead to the same behaviour. This time varies with the hardenability of the steel and is characteristic of it. It is therefore called by Grossmann the "characteristic cooling time." In a round steel bar, at a certain point—the radial distance of which from the surface varies with the composition, etc., of the steel—the structure is 50% martensite, 50% troostite, representing a half-hardened condition. The time taken *at this point* to cool (during quenching) to a temperature half-way between that of the quenching temperature and that of the quenching medium, is called the "characteristic cooling time."
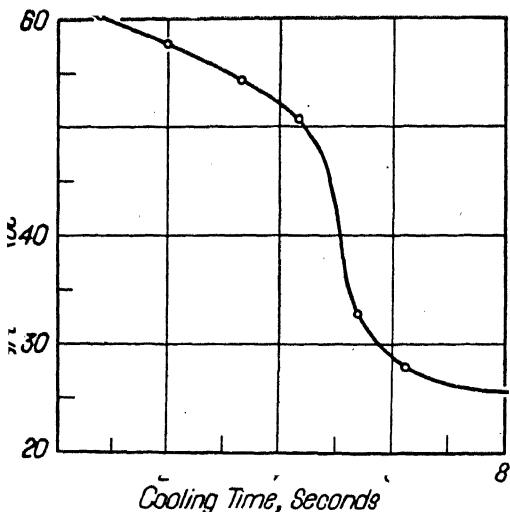
FIG. 44. HARDNESS FALLS OFF AS COOLING TIME INCREASES FROM SURFACE TO CENTRE OF A 1-IN. ROUND STEEL BAR. (*Grossmann.*)

In a quenched steel of a given composition, the cooling time at the centre increases with the size of the bar, and with a certain size it will coincide with the "characteristic cooling time."

In Fig. 45 the hardness-penetration diagrams of a series of round bars (varying from ⅝ in. to 1½ in. diameter) of a carbon-molybdenum steel are given. The steel had the following composition, viz. :—

Carbon	Manganese	Phosphorus	Sulphur	Silicon	Molybdenum
0·66%	0·76%	0·013%	0·026%	0·27%	0·22%

When the hardness values at the centres of these bars are plotted against diameters of the bars, the curve A (Fig. 46) is obtained. Note the sharp fall which occurs when the diameter just exceeds 1 in. The "characteristic cooling time " is assigned to the centre of a bar of 1 1/32 in. diameter.

In such a series, for which the quenching severity was the same throughout, the bar in which the characteristic cooling time occurs at the centre, is regarded as the " critical size " and is taken as an index of the hardenability of the steel. The " critical size " of a steel of greater hardenability

will be larger, *e.g.* that of a steel of the following composition, viz. :—

Per cent.

Carbon	Manganese	Phosphorus	Sulphur	Silicon	Molybdenum
0·70	0·86	0·021	0·024	0·27	0·24

Curve B (Fig. 46) was plotted from a set of bars of this steel similarly graded in size to those of curve A and similarly quenched. The critical size for this steel is about $1\frac{1}{4}$ in. diameter.

Another way of finding the critical size is to ascertain in which bar the structure is 50% martensite or troostite (corresponding with the abrupt change in hardness) at the centre. Any bar of larger diameter will show a structure of less than 50% martensite at the centre, i.e. an unhardened core. Hence the critical size may be designated as the largest size which contains no unhardened core.

These relationships are shown in the case of a straight carbon steel of the following composition, viz. :—

Per cent.

Carbon	Manganese	Phosphorus	Sulphur	Silicon
0·89	0·78	0·011	0·024	0·22

The hardness-penetration curves of the series are given in Fig. 47, providing the data from which the " critical size" curve of Fig. 48 is plotted. The most abrupt fall occurs at a diameter of 1·05 in., this therefore being taken as the critical size. By examination of etched cross-sections of the bars it was found that an unhardened core existed in the $1\frac{1}{8}$ in. size but not in the 1 in. size, the critical size therefore being in between. This roughly verifies the result obtained by the first method (Fig. 48). Although a bar may be hardened throughout, as judged by a fracture or etching test, it is not necessarily at *full* hardness throughout its section, *e.g.* the 1 in. bar of
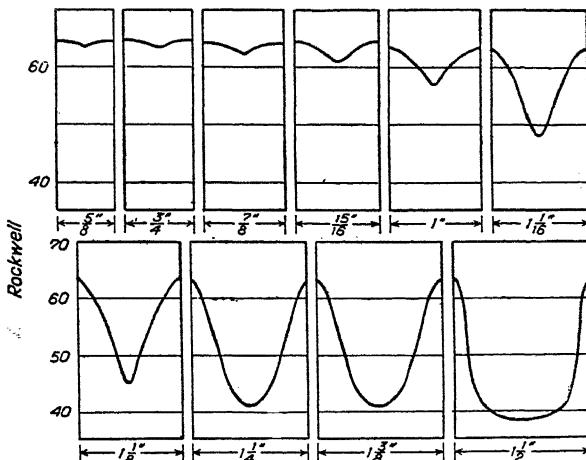
Fig. 45. Hardness Penetrations in Steel Bars of increasing Diameter. Composition : C 0·66%, Mo 0·22%, Mn 0·76%.
(*Grossmann.*)

85
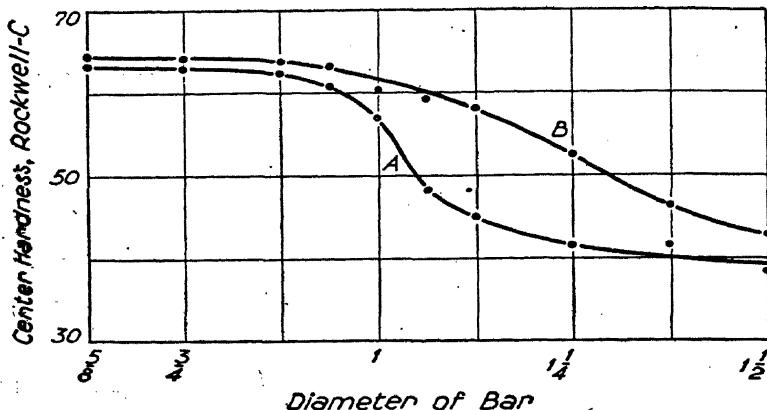
FIG. 46. CENTRE-HARDNESS CURVE OF QUENCHED STEEL BARS OF
INCREASING DIAMETERS.

CURVE A IS FROM FIG. 45.
CURVE B IS FROM A STEEL CONTAINING C 0·70 %, Mo 0·24 %, Mn 0·86%.

(*Grossmann.*)
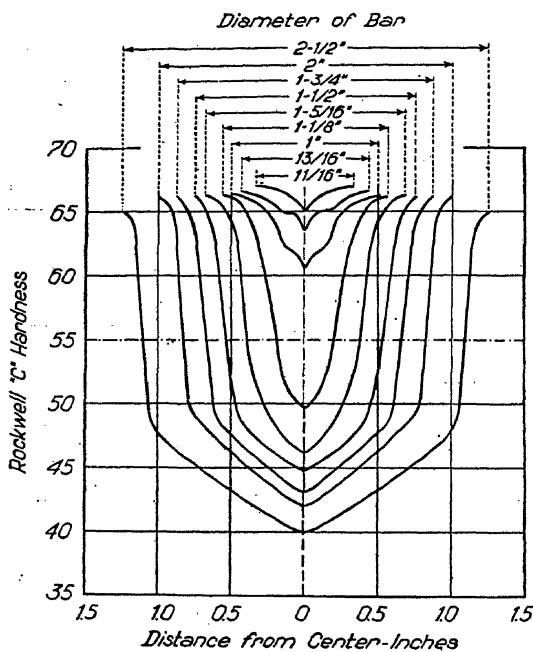
FIG. 47. HARDNESS DISTRIBUTIONS IN A SERIES OF SIZES OF ROUND
BARS OF 0·89% CARBON STEEL, WATER-QUENCHED. QUENCHING
SEVERITY (H) = 1·49. (*Grossmann.*)

86
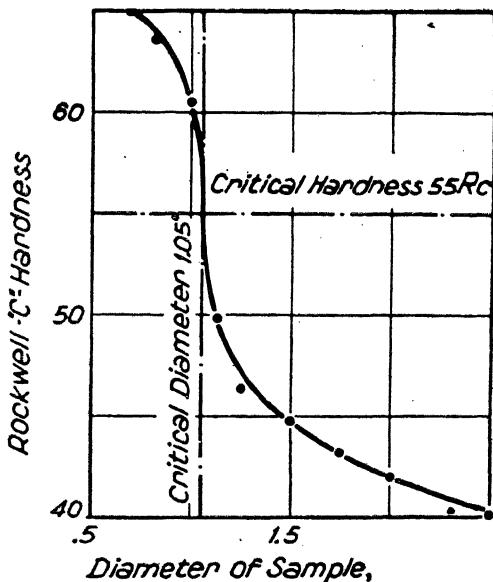
FIG. 48.   CENTRE-HARDNESS CURVE OF QUENCHED STEEL (S.A.E. 1090).
C 0·89%, MN 0·78%.   DATA FROM FIG. 47.

Fig. 47, although showing no unhardened core when etched, has a distinct hardness gradient between edge and centre.

The response of high-carbon steels to heat-treatment varies widely, even if the steels are of the same composition as determined by the usual routine chemical analysis.   Bain* has published hardness-penetration curves on eight 1 in. diameter bars of 1% carbon steel, all given the same hardening
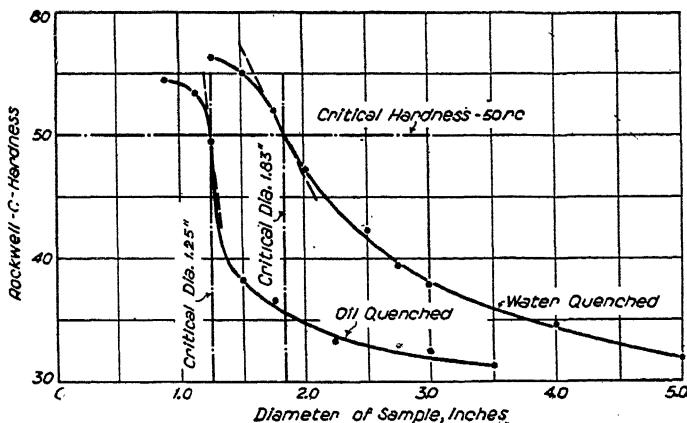
FIG. 49.   CENTRE-HARDNESSES OF QUENCHED ROUND STEEL BARS,
PROGRESSIVELY INCREASING IN DIAMETER.
ONE SERIES OIL-QUENCHED.   ONE SERIES WATER-QUENCHED.

* E. C. Bain.   Trans. Amer. Soc. Steel Treating.   1932, 20, 385–428.

treatment. Only one of the eight bars showed full hardening throug the section, the hardness at the centre of which was 63 Rockwell C. The centre hardness-values of the remaining seven bars varied between 50 and 40. The chief factors regarded as responsible for this variable hardenability are oxides and grain size.

of Quenching.

The hardening of a piece of steel depends not only on its hardenability, but also on the severity with which it is quenched. Therefore a steel of a given composition will exhibit different critical sizes for different degrees of severity of quenching.

For example, in Fig. 49 the critical size of a nickel-chromium steel, when oil-quenched, is 1·25 in., but when water-quenched it is 1·83 in. The composition of this steel was as follows :—

Per cent.

Carbon	Manganese	Phosphorus	Sulphur	Silicon	Nickel	Chromium
0·41	0·76	0·017	0·033	0·18	1·20	0·67
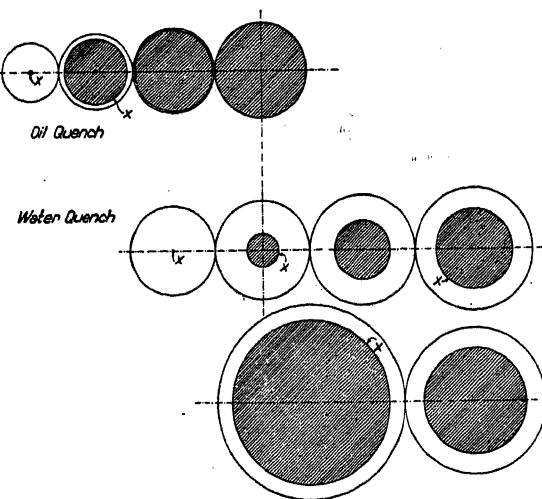
FIG. 50. DIAGRAMMATIC REPRESENTATION OF ETCHED CROSS-SECTIONS OF STEEL (S.A.E. 3140), QUENCHED (TOP SERIES) IN OIL, (BOTTOM SERIES) IN WATER. COMPOSITION OF STEEL : C 0·41%, MN 0·76%, NI 1·20%, CR 0·67%.

In Fig. 50 the behaviour of the two sets of bars is shown diagrammatically. The shaded areas represent the unhardened cores. The deeper penetration of the hardening effect of water-quenching is shown in the larger critical size (1·83 in. diameter) for water-quenching as compared with that for oil-quenching (1·25 in.). As the size increases beyond the critical size it will be noticed that the proportion of unhardened core to hardened rim increases much more rapidly in the oil-quenched specimens than in the water-quenched specimens, in which the hardened rim, although diminished in thickness, is still present in quite large sizes.

The quenching severity of various methods is called by Grossmann the H-value. A table of H-values is given below (Table X) involving various quenching media and different degrees of agitation of the liquid or of the piece being quenched.

TABLE X.

Table of H-values. (Quenching Severity.)

Mode of Cooling.	Air.	Oil.	Water.	Brine.
Specimen and coolant still	0·02	0·30	1·0	2·2
Specimen moved moderately, coolant still		0·4 to 0·6	1·5 to 3·0	
Specimen moved vigorously, coolant still		0·6 to 0·8	3·0 to 6·0	7·5
Strong to violent current or spray of coolant		1·0 to 1·7	6·0 to 12·0	
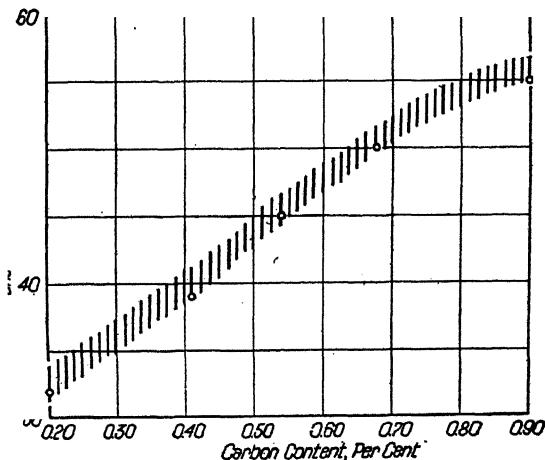
FIG. 51. INFLUENCE OF CARBON ON HARDNESS OF SPECIMENS CONTAINING, AS QUENCHED, 50% MARTENSITE.

Important facts revealed by the above figures are :—

(a) Quenching severity is greatly affected by the degree of circulation.

(b) Quenching severity in greatly agitated oil may be greater than in still water.

Carbon Content.

In some of the hardness-penetration curves (*e.g.* Fig. 40) the exact position of the 50% martensite structure is not clearly shown owing to the absence of any sharp inflection in the curve. If the hardness of this 50% martensite structure were known, its position, and consequently the depth of hardening penetration, could be more definitely ascertained.

As the hardness of martensite varies with the composition (mainly carbon content) the hardness of the 50% martensite zone will vary similarly.

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In Fig. 51 this variation is illustrated by a band of values assembled from available data. These are working values which are apparently quite close to reality for all the common carbon steels, though the hardnesses are likely to be higher in alloy steels.
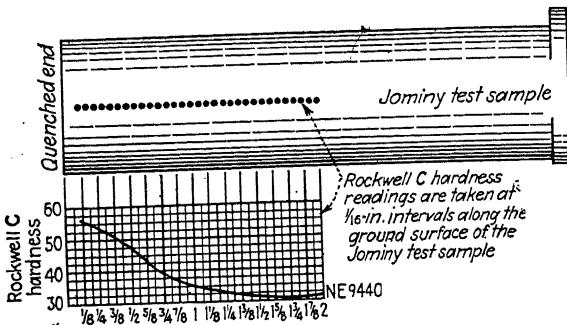
FIG. 52. DIAGRAM SHOWING LOCATION OF HARDNESS DETERMINATIONS, ON A JOMINY, END-QUENCH HARDENABILITY TEST SAMPLE. NE 9440 STEEL.

Jominy Test for Hardenability.

In the Jominy end-quench test it is possible to determine with a single test how a steel will harden at different cooling rates between about 500° C. per sec. and 2° C. per sec. This test has been described by J. Winning[*] and others.

The test piece used for testing the hardenability of U.S. National Emergency (NE) steels under the direction of the American Iron and Steel Institute is obtained by machining a sample of the steel to 1-in. round, $3\frac{7}{8}$ in. long, with a flange on one end, $1\frac{1}{4}$ in. diameter and $\frac{1}{8}$ in. thick (Fig. 53). The sample is first normalized, machined to size, and then heated to the quenching temperature appropriate to its composition. It is then inserted

(a) The test-piece : a simple cylinder with flanged end.

(b) (right).—The quenching fixture, made of welded strip with hole for test-piece.

(c) (extreme right).—Quenching fixture conveniently set over a tank containing the quenching liquid.
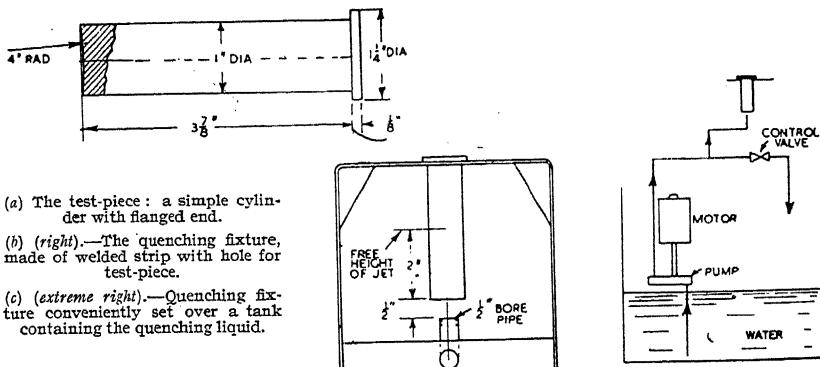
FIG. 53. SPECIMEN AND APPARATUS USED IN THE JOMINY END-QUENCH HARDENABILITY TEST. (Winning.)

' Mech. World." 1943, 113, pp. 573-575.

in a hole in a fixture (see Fig. 53) so that it hangs vertically by the flange over a $\frac{1}{2}$-in. round orifice which is $\frac{1}{2}$ in. below the bottom end of the sample. Water from the orifice rises with sufficient pressure to reach a height of $2\frac{1}{2}$ in. when the sample is not in place, this height being controlled by a gate-valve (Fig. 54). When making the test, the water is allowed to flow until the sample is practically cold. Smooth, flat surfaces are then ground 0·015 in. deep, lengthwise, on diametrically opposite sides of the sample. Rockwell hardness values are then obtained at $\frac{1}{16}$ in. intervals from the quenched end, along the length of the flat surfaces as shown in Fig. 52.

The rate of quenching is very fast on the bottom end, and as the heat must pass through the sample by conduction, the rate diminishes vertically upwards, being very slow in the top end of the test piece.

Thus, in one piece of steel, is obtained a quenching-rate gradient from very fast to very slow, and the Rockwell hardness tests give a corresponding hardness gradient. Different steels will show different hardness gradients for a standard quenching-rate gradient. Interpretations of this test after determination of tensile strength, yield point, reduction of area and elongation will apply only to fully-quenched steels ranging in quenched hardness from approximately 200 to 400 Brinell. These interpretations do not apply to extremely hard steels which have been treated to give a hardness value below 200 Brinell, nor to steel that is annealed, normalized, or in the as-rolled condition.

In the original test devised by Jominy and Boegehold, the test piece was first carburized before being heated for the end-quench test. Direct-hardening steel test pieces are not carburized. Steels of low hardenability harden only to a slight depth from the hardened end, making precise measuring difficult. A special form of test bar (illustrated on right of Fig. 55) was adopted. In this form the hardening effect penetrates for a greater distance and is thus measured with greater accuracy, the data being compared with the data from a standard bar.

Additional data in connection with the standard Jominy test are as follows :— Temperature of water .. 24° ± 2·8° C. (75°±5° F.)

Specimens shall be machined from a bar previously normalized and of such a size as to permit the removal of all decarburized metal in machining to 1 in. round. The end of the specimen to be quenched shall have a reasonably smooth finish, preferably produced by grinding. (The normalizing treatment may be omitted by agreement between seller and purchaser.)

The test specimen should be heated, within a period of thirty to forty minutes, to the appropriate hardening temperature, at which it should be maintained for twenty minutes before quenching.

It is necessary to take rigorous precautions to prevent scaling and decarburization. The specimen may be heated, in a vertical position, quenching-end resting on a layer of clean cast-iron swarf at the bottom of a container, provided with an easily removable cover and with a thermocouple for temperature-control and trimming. The stream of water shall be directed against the bottom face of the specimen for not less than ten minutes, and a condition of still air maintained around the specimen during that period. If not cold by then it shall be removed and quenched immediately in cold water.

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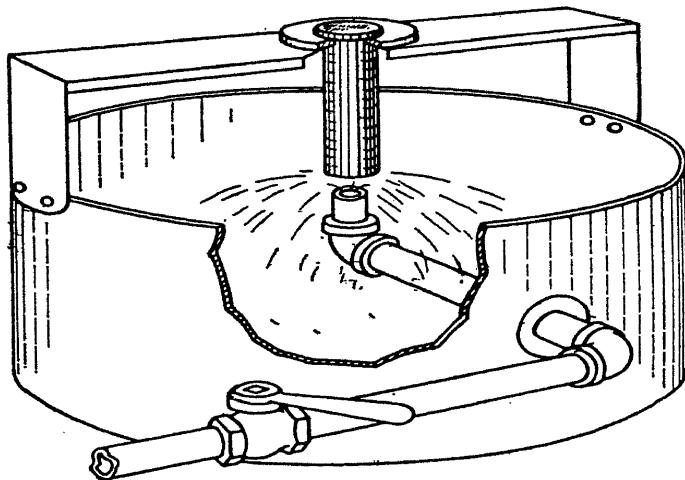
FIG. 54. CUT-AWAY VIEW, SHOWING FIXTURE FOR MAKING JOMINY HARDENABILITY TESTS.
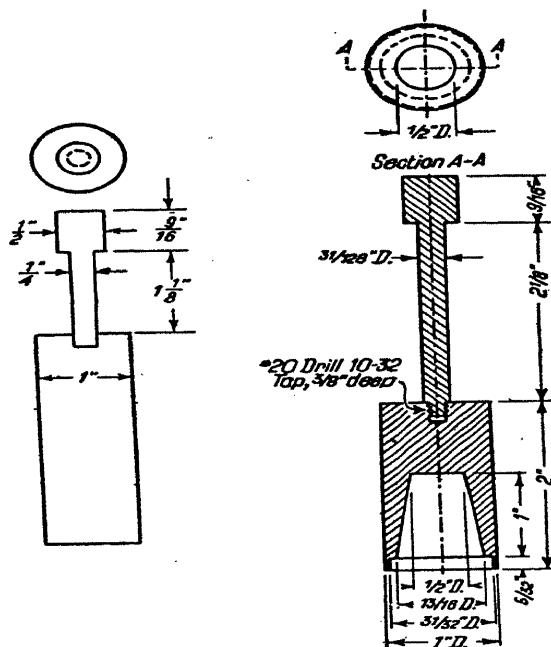
FIG. 55. SPECIAL FORM OF TEST-PIECE FOR JOMINY TEST ON STEELS OF LOW HARDENABILITY (RIGHT). STANDARD FORM (LEFT).

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Winning* suggests a jig, suitable for holding the Jominy specimens when making hardness tests on a Vickers hardness-measuring machine. It consists of a movable Vee-block on which the sample may be changed in position, being made adjustable by a screw working against a compression spring (Fig. 56). The screw may be a bolt, screwed 16 threads per inch, with one face on the hexagon head marked so that each complete turn of the screw advances the sample $\frac{1}{16}$ in. If the jig is set up so that the centre of the diamond indenter is exactly on the end of the test-bar when clamped in position on the block and the marked face of the hexagon up, then one complete turn of the screw to bring the marked face up once more will advance the sample by $\frac{1}{16}$ in. for a fresh reading. In this way a series of equidistant readings is obtained, suitable for the preparation of a graph.
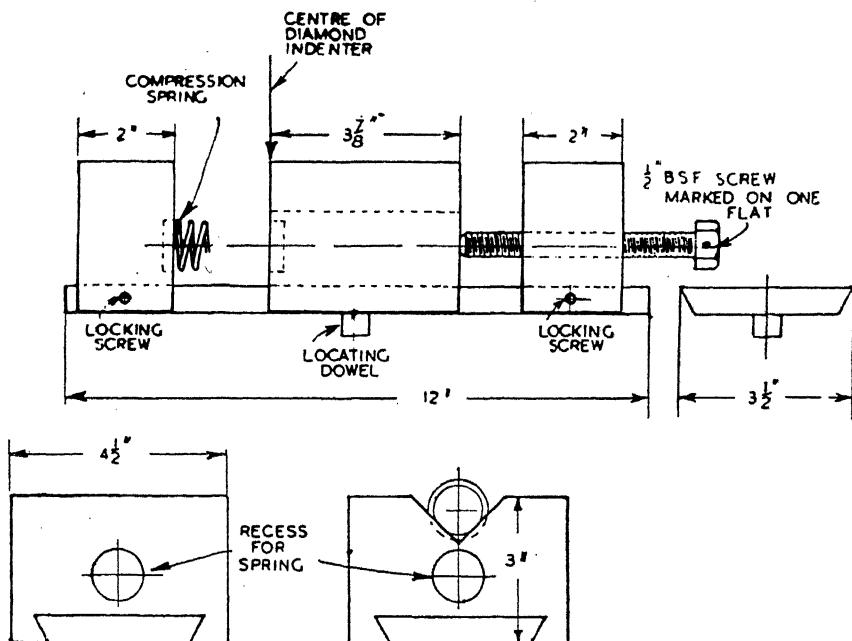
FIG. 56. A SUGGESTED FORM OF JIG TO SYSTEMATIZE THE DETERMINATION OF HARDNESS ON JOMINY TEST-PIECES, IN THE VICKERS HARDNESS TESTER.
(*J. Winning.*)

Practical Applications of Jominy Test-results.

The usefulness of any test is greatly enhanced if from its results, other properties may be evaluated, thus saving much time and expense in the preparation of special test-pieces for direct determination. An end-quench test is relatively simple and inexpensive.

The deductions which may be made from the result of a Jominy test are numerous and useful. Careful laboratory experiments have indicated just how rapidly a steel test-piece cools at different distances from the quenched end. These rates of cooling are given in Table XII.

* " *Mech. World.*" *1943. 113, 573–575.*

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TABLE XII.

Cooling Rate in Jominy Test.

Rate of cooling in degrees Fahrenheit per sec. (at 1300° F.)
at the distance indicated from the water-cooled end of a
standard Jominy end-quenched specimen, in sixteenths of
an inch.

$\frac{1}{16}$ in. — 600° F./sec.	$\frac{7}{16}$ in. — 37° F./sec.	1 in. — 10° F./sec.
$\frac{1}{8}$ in. — 190 ,,	$\frac{1}{2}$ in. — 30 ,,	$1\frac{1}{8}$ in. — 8·5 ,,
$\frac{3}{16}$ in. — 99 ,,	$\frac{9}{16}$ in. — 26 ,,	$1\frac{1}{4}$ in. — 7·2 ,,
$\frac{1}{4}$ in. — 72 ,,	$\frac{5}{8}$ in. — 22 ,,	$1\frac{1}{2}$ in. — 5·5 ,,
$\frac{5}{16}$ in. — 56 ,,	$\frac{3}{4}$ in. — 18 ,,	2 in. — 4·3 ,,
$\frac{3}{8}$ in. — 44 ,,	$\frac{7}{8}$ in. — 14 ,,	

TABLE XIII.

Cooling Rates of Round Bars.

Rate of cooling in degrees Fahrenheit per sec. (at 1300° F.)
at surface, half-radius and centre of different-sized rounds—
quenched in water and oil.

Diameter " D."		Surface.	Half-radius.	Centre.
		Deg. F./sec.	Deg. F/sec.	Deg. F./sec.
1-in. Round—				
Water-quenched	..	850	135	100
Oil-quenched ..	..	120	53	45
2-in. Round—				
Water-quenched	..	550	46	32
Oil-quenched ..	..	58	24	18 ·
3-in. Round—				
Water-quenched	..	400	27	15
Oil-quenched ..	..	30	12	9
4-in. Round—				
Water-quenched	..	100	14	8
Oil-quenched ..	..	15	$6\frac{1}{2}$	$5\frac{1}{4}$

Table XIII indicates the cooling rate at the surface, half-radius and
centre of different-sized bars when water-quenched and also when oil-
quenched. These figures were also obtained by laboratory experiments.

If it is desired to know how hard a certain type of steel will be at the
centre of a water-quenched 3-in. round bar, for example, this may be
approximately determined from the Jominy test of the same steel.

Table XIII indicates that a 3-in. round bar, when quenched in water,
will cool at the rate of 400° F. per sec. on the surface, 27° F. per sec. at the
half radius, and 15° F. per sec. at the centre.

Table XII shows that the Jominy test-piece cools at 15° F. per sec. at
approximately $\frac{7}{8}$-in. from the quenched end, and, therefore, its hardness
at this point will be practically the same as that of the 3-in. round bar
at its centre, owing to equality in cooling rates.

Suppose that it is desired to know the *approximate* value of tensile strength, yield point, elongation and reduction of area of this same 3-in. round bar after quenching.

The method of working out is best shown by taking an actual case.

Assume that a Jominy test has been made on a sample of NE 9440* steel. It is required to determine what properties will be exhibited at half-radius on a 3-in. round bar of this steel when oil-quenched and before tempering. The following procedure will give the required information :—

(1) Table XIII shows that a 3-in. round, oil-quenched specimen will cool at the rate of 12° F. per sec. at half-radius.

(2) Table XII shows that the Jominy test specimens cool at approximately 12° F. per sec. at a point $\frac{15}{16}$-in. from the quenched end.

(3) The sketch, Fig. 52, shows that at $\frac{15}{16}$-in. from the quenched end of the Jominy test of the NE 9440 sample the hardness was 35 C. Rockwell. It is therefore known that the hardness of the 3-in. round at half-radius will be 35 C. Rockwell, because at that point the hardening rate was the same as that of the Jominy test at $\frac{15}{16}$-in. from the quenched end.

(4) Conversion Table XIV shows that 35 C. Rockwell is approximately the same as 325 Brinell.

(5) Table XIV shows the tensile properties, corresponding to 325 Brinell, to be as follows :—

Yield Point. Tons/in.²	Tensile Strength. Tons/in.²	Elongation. % on 2 in.	Reduction of Area %
61–65	68–76·5	16–20	47–56

Therefore, these properties will be approximately those possessed by the 3-in. round bar at half-radius in the oil-quenched condition.

TABLE XIV.

Approximate tensile properties in relation to hardness.

Rockwell C. Hardness.	Brinell Hardness.	Yield Point. Tons/in.²	Tensile Strength. Tons/in.²	Elongation. % on 2-in.	% Reduction of Area.
15	200	31·0–35·5	42·5–47·0	22/28	60/68
20	225	38·5–82·0	47·0–53·0	21/26	57/65
24	250	45·5–50·0	52·5–59·0	20/25	55/63
28	275	51·5–55·5	58·0–64·5	18/23	52/61
31	300	56·5–60·0	63·5–70·5	17/21	50/58
34	325	61·0–65·0	68·0–76·5	16/20	47/56
37	350	67·0–71·0	73·5–82·5	14/18	45/54
39	375	72·5–76·5	78·5–89·0	13/16	42/51
42	400	78·0–82·0	74·5–98·5	11/15	40/49

The curves in Fig. 57 are taken from the article by Winning to illustrate the difference in hardenability of two steels of different manganese content. The quenching temperature was 920° C. for both steels.

* *Approximate composition, C. 0·40%, Mn. 1·0%, Si. 0·5%, Ni. 0·35%, Cr. 0·35%, Mo. 0·11%.*

Supposing it is required to know whether steel A would fully harden to the centre when water-quenched. From Table XIII the cooling rate at the centre of a 1-in. round bar, water-quenched, is found to be 100° F. per sec. The position, on a Jominy test-bar, of this rate of cooling, is about $\frac{3}{16}$-in. from the quenched end,* and this position, on the Jominy test-bar of steel A, corresponds to a V.P.N. of 450. Even at the quenched end itself the hardness number is less than 600, and this steel, therefore, is quite unsuitable. In the case of steel B, however, the same rate of cooling
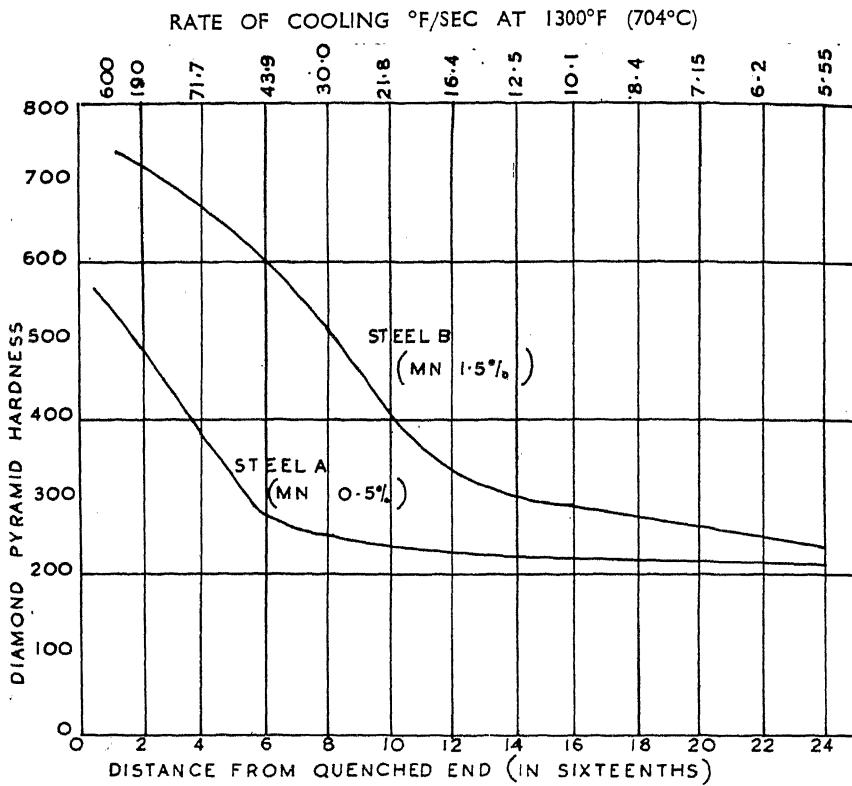
FIG. 57. END-QUENCH HARDENABILITY CURVES OF 0·3 % C STEELS OF DIFFERENT MANGANESE CONTENT.

will be obtained at the centre of a 1-in. round bar, viz., 100° F. per sec., and the hardness corresponding to this rate, on a Jominy test-piece of this steel, is over 700 V.P.N. If this steel, however, will fail by cracking, under such severe quenching treatment (water-quenching from 920° C.), it would be desirable to ascertain whether full hardening to the centre could be obtained by oil-quenching. The rate of cooling for a 1-in. diameter bar at the centre, when oil-quenched →(Table XIII) is 45° F. per sec., and this rate of cooling occurs at about $\frac{6}{16}$-in. from the quenched end in the Jominy

* See Table XII.

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test, Table XII. The corresponding hardness at this point on a Jominy test-bar of this steel is 600 V.P.N., Fig. 57. Thus full-hardening to the centre could be obtained by oil-quenching 1-in. rounds of steel B. Full-hardening to the centre of larger bars, e.g. 1¼-in. diameter, could be obtained by water-quenching, and to ascertain if this steel could be safely hardened by water-quenching from a temperature lower than 920° C., the possibilities could be explored by further tests of the same kind.

The Jominy test may be used to determine the hardenability of steels of any composition, especially new steels, and to deduce therefrom the physical and mechanical properties, e.g. oil-toughened heavy axles, specified to possess a minimum tensile strength in the centre, may be made from a steel, the Jominy test on which has shown that such strength can be obtained by suitable heat-treatment. Tables are available (see Table XIV) which give reliable conversions from hardness values to approximate values of tensile strength. The data provided in the Jominy test procedure indicate whether such values may or may not be realised. In no other way could so much information be obtained in so short a time, with such simple equipment and at so low a cost.

Roush* has provided data which show that a standard steel used for gears could be replaced by National Emergency (U.S.A.) steels. The standard steel gave a certain hardness in the centre at the root of a gear tooth, and hardenability curves showed that the N.E. steels had the same hardness at the same point in the same gear.

Fig. 58 shows curves for deep-hardening grades. The hardenability curve of the standard steel (SAE 4340) has been matched by manganese-molybdenum and chromium-nickel-molybdenum steels, but not by the carbon-molybdenum steel. The compositions of the four steels were as follows :—

TABLE XVI.

Compositions of Steels of Fig. 58.

Type.	Per cent.					
	Carbon.	Manganese.	Silicon.	Chromium.	Nickel.	Molyb-denum.
S.A.E. 4340 ..	0·41	0·64	0·24	0·75	1·78	0·19
Mn.—Mo. ..	0·43	1·60	0·31	—	—	0·37
Cr.—Ni.—Mo. ..	0·46	1·17	0·19	9·50	0·50	0·33
C.—Mo. ..	0·73	0·89	0·28	0·20	0·14	0·25

These deep-hardening grades are used for heavy sections where other steels will not harden through.

It is necessary to allow commercial limits in any steel, and it cannot be expected that hardenability can be matched exactly to the line. By performance-testing it has been shown, for instance, that a given gear will be satisfactory with a core hardness range of, say, 10 points Rockwell C.

In specifications for hardenability, likewise, limits have to be considered.

* *loc. cit.*
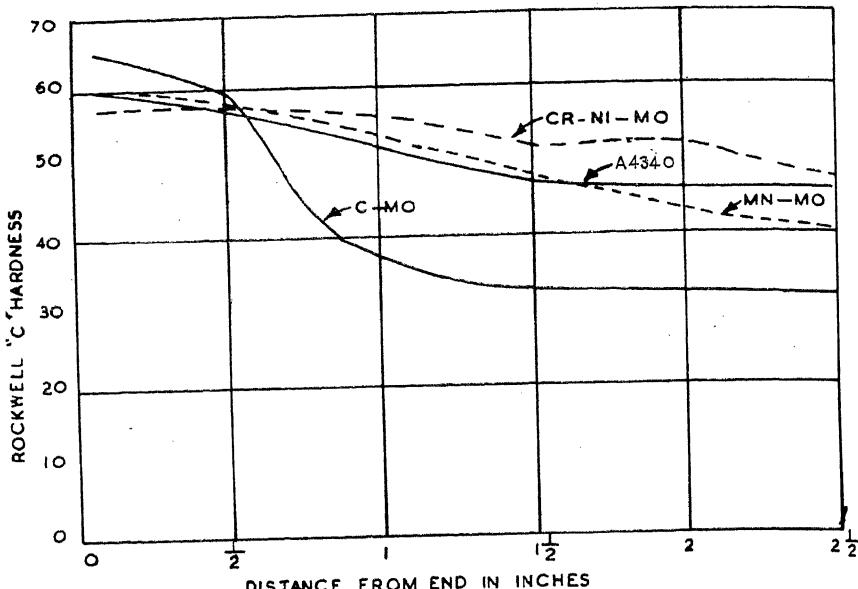
FIG. 58. STANDARD END-QUENCH HARDENABILITY CURVES FOR VARIOUS STEELS.
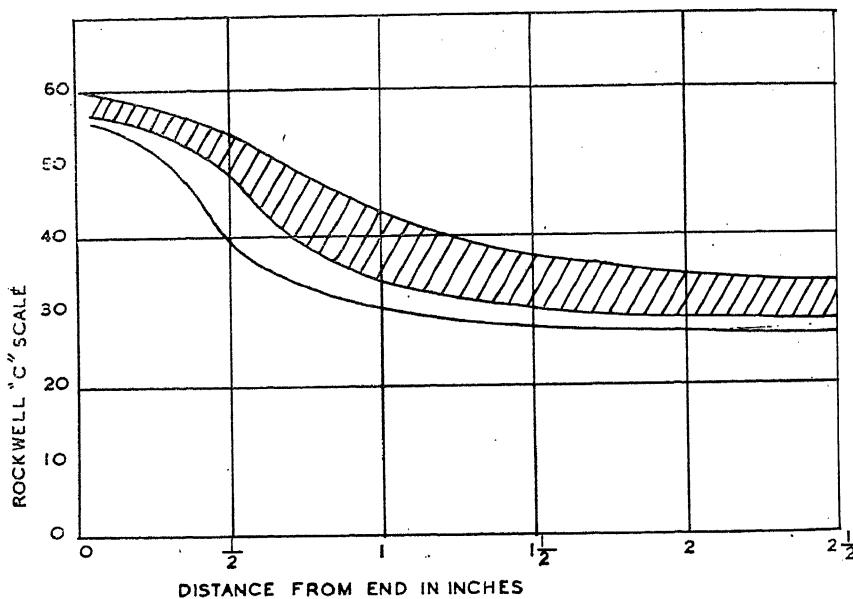
FIG. 59. BAND OF ACCEPTABLE TESTS FOR HARDENABILITY OF NE 8744 STEEL. LOWER CURVE REPRESENTS A STEEL BELOW STANDARD.

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A hardenability band is therefore established instead of a single-line curve. One example of this is shown in Fig. 59. The hardenability band shows the limits required for axle drive shafts. The curve below the band represents a " heat " which would not give the required hardness in production.

Other Tests for Hardenability.

(1) *Cross-Sectional Hardness Tests.*

The method of determining hardness penetration by obtaining hardness-values on a ground and polished face of a cross-section of a hardened round steel rod has already been described (p. 81, Fig. 39). By plotting hardness values, determined at regular intervals along the diameter, against distances between edge and centre, the diminution in hardness radially inwards is clearly shown, also the fall in hardness associated with increase in diameter of rod. The results obtained vary with the severity of quenching. Rods of known hardenability can be used as a check on quenching conditions or a sample rod from a batch can be treated and the results used as an acceptance test.

(2) *Shepherd P.F. Test.*

One piece of a rod, quenched under standard conditions, and fractured, is ground and *polished* to give a circular cross-section. This face is etched in order to reveal, by the selective action of the etching reagent, the depth of the outer hardened zone. The occurrence of soft spots is revealed by this, the P part of the test.

The other piece provides the F part of the test, the *fractured* surface being compared with a series of standards, thus serving as a rapid guide to the grain size of the steel.

This test, used in the tool trade, has limited application as a general test.

(3) *Burns and Riegels Test* (S.A.C. test).

This test is applied to steels down to $0 \cdot 1\%$ carbon. Test-pieces, quenched under standard conditions, are fractured and the surface is prepared for a hardness traverse along a diameter. The following data serve as an indication of hardenability, viz. :—

$$S \; = \; \text{surface hardness.}$$
$$A \; = \; \text{average hardness.}$$
$$C \; = \; \text{centre hardness.}$$

Data Predictable from Jominy Test Results.

For more accurate, quantitative determination of hardening depth, the data provided by the Jominy test and by the cross-sectional hardness penetration test are utilised. J. L. Lamont[*] has fully described this method, as an aid in predicting the response of emergency alloy steels to hardening. The charts used by Lamont enable one to predict, from the results of a Jominy end-quench hardenability test :—

(1) The severity of quench—if, in addition to the Jominy test data, the cross-sectional hardness distribution is known, of a single bar of the same steel, quenched under the desired quenching conditions;

(2) The diameter of round bars having a desired hardness at any point between the centre and surface, for a predetermined severity of quench;

[*] " *The Iron Age.*" *Oct. 14 1943. pp. 64-70.*

(3) The cross-sectional hardness distribution in a round bar of a given size for a predetermined severity of quench ;

(4) The dimension of square bars and 1 : 2 flat bars and the thickness of plates having a desired hardness at the centre for a predetermined severity of quench ;

(5) The hardness at the centre of square bars, 1 : 2 flat bars, and plate for a predetermined severity of quench.

Round Bars.

The degree of hardening that is attained, by quenching, at a given point in a bar, is dependent not only on the inherent hardenability characristics of the steel and the severity of the quench, but also on the actual rate resulting from the severity of quench, bar size, bar shape and position within the bar. For purposes of mathematical treatment, the severity of quench is evaluated by the ability of the quenching medium to lower the surface temperature of the bar. It is assumed that the temperature gradient within the bar is controlled by the surface temperature. Thus the greatest conceivable severity of quench is when the surface is cooled instantaneously to the temperature of the quenching medium. Lesser severity of quench and larger sections lead to less rapid lowering of surface temperature and less steep temperature gradients between surface and centre. These relations have been calculated by Russell,* so that it is possible to estimate the time required for a given fall in temperature to occur at any point in a round, square, flat, or plate section during quenching. It is not accurately known what is the exact temperature—critical in the hardening of steel—intermediate between that of the " quench " and the quenching medium. Asimow, Craig and Grossmann,† however, have determined the times required for positions along a Jominy end-quenched specimen to cool to half the temperature difference between the quenching temperature of the bar and the temperature of the quenching medium. For convenience, therefore, Lamont has used the "half-temperature" time in constructing his charts from Russell's constants, using a thermal diffusivity coefficient of 0·009 sq. in. per sec. Grossmann's " H " values are used to express severity of quench, instead of Russell's " h " (which, it should be noted, is equal to 2 " H ").

The depth of hardening, which may be measured in terms of microstructure, or more conveniently as hardness, is expressed as the ratic $\frac{r}{R}$, where r is the radius of the unhardened core and R is the radius of the bar when fully hardened.

This fractional distance $\frac{r}{R}$ is zero at the centre and 1·0 at the surface of the bar when fully hardened.

The severity of quench is markedly influenced by rate of agitation composition, viscosity, and other properties of the quenching medium surface scale, etc.

* T. F. Russell, " Some Mathematical Considerations on the Heating and Cooling of Steel." I.S.I. Specia Report, No. 14, 1936, 149–187.
† " S.A.E. Trans." 1941, 36–49, pp. 283–292.

In order, therefore, to apply the data derived from the Jominy test-bar to actual quenching conditions, the severity of quench for a given quenching condition should be determined. The range of quenching severities obtainable with one and the same medium under different conditions is shown in Table X. The fastest possible quench, which results in the surface temperature of the quenched bar being instantly lowered to that of the quenching medium, has been designated the *ideal quench*. The severity of an ideal quench is expressed as " H " = ∞.
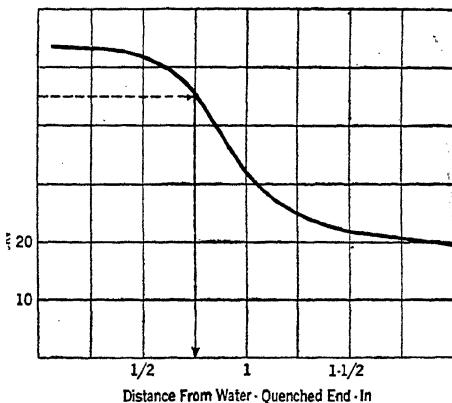
FIG. 60. JOMINY HARDENABILITY OF A STEEL SAMPLE.

Severity of quench may be established by direct measurement of cooling rate, but it is more convenient to measure it as a function of the depth to which a bar of steel hardens. This may be accomplished by hardening bars of different sizes (Grossmann, Asimow and Urban*), but Lamont, as already pointed out, based his method on the hardness data obtained from the Jominy hardenability test and from cross-sectional hardness data of one bar hardened incompletely at the centre.

In order to establish a relation between cooling rate (severity of quench, bar size, bar shape and position in bar) and hardenability, it is necessary to assume that a given cooling rate, when applied to the same steel, will produce the same hardness, regardless of bar size, position in the bar, or quenching medium used. That is, a specific " half-temperature " time will always produce the same hardness in the same steel at the position in the bar at which that " half-temperature " time is attained.

With this assumption, the charts can be used for estimating the *severity of quench* from Jominy test data and cross-sectional hardness values of a quenched bar of the same steel. For example, assume that a steel has a Jominy hardenability similar to that illustrated in Fig. 60 and cross-sectional hardness values as shown in Fig. 61 after quenching a 4 in. diam. round bar. Then, selecting some convenient hardness, say, 45 Rockwell C, it will be seen from Figs. 60 and 61 that 45 Rockwell C hardness occurs at a depth

* *A.S.M. Publication, " Hardenability of Alloy Steels." 1939, pp. 124–190.*

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of $\frac{3}{4}$ in. from the water-cooled end of the Jominy bar and also at a distance of 1 in. from the centre of the 4 in. diameter bar at a fractional distance of $0\cdot5$; that is $\frac{r}{R} = 0\cdot5$.

In Fig. 62 is a diagram which shows hardening for a fractional distance of $\frac{r}{R} = 0\cdot5$, and the point at which the line indicative of $\frac{3}{4}$ in. on the Jominy bar intersects the line representative of a 4 in. diameter bar falls on the curve signifying a quenching severity of $H = 2\cdot0$.
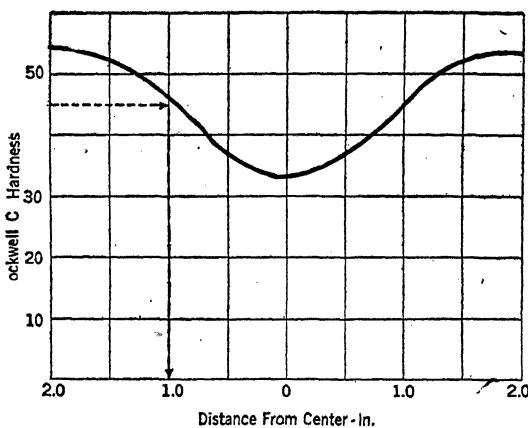
FIG. 61.   CROSS-SECTIONAL HARDNESS OF A STEEL SAMPLE.

Once the H value for severity of quench has been established on one steel for a given quenching condition, this H value remains the same as long as quenching conditions remain constant. It is, therefore, possible to use this H value in order to estimate the depth of hardening of another section or another steel, provided that the quenching conditions are unchanged.

Reference to the Jominy hardenability curve, the fractional distance curves and the table of quenching severities (Table X) serves as a guide to the selection of the size of bar on which to determine cross-sectional hardness for establishing severity of quench values.

For example, suppose that the greatest change in the slope of the Jominy hardenability curve (a convenient location for selection of a hardness value) occurs at about $\frac{1}{2}$ in. from the water-cooled end, and an agitated oil quench is to be used. It will be observed from the table that H values of from $0\cdot6$ to $0\cdot8$ are obtained with this type of quench. Then by reference to the various charts available, it will be seen that with this type of quench, a round bar about 2 in. diameter would harden at a fractional distance of $\frac{r}{R} = 0\cdot3$, to the same degree as would the Jominy bar at a distance of $\frac{1}{2}$ in. from the water-cooled end.

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A smaller size bar might harden too deeply and a much larger bar would be uneconomical.

Depth of Hardening.

It is often desirable to predict the degree of hardening to be expected at specific distances from the surface of quenched round bars.

If optimum mechanical properties are to be produced in a steel bar, it is probable that, for a certain severity of quench, its diameter should not exceed that which will harden at the centre, to a hardness represented by the " shoulder " corresponding e.g. to $\frac{9}{16}$ in. from the quenched end of a Jominy hardenability specimen (Fig. 60).
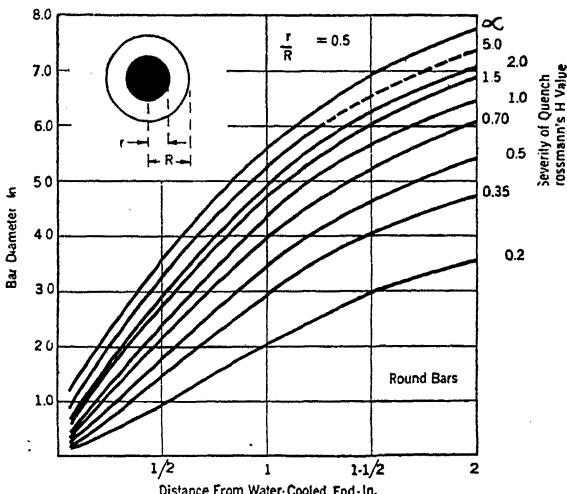
FIG. 62. LOCATION ON END-QUENCHED JOMINY HARDENABILITY SPECIMEN, CORRESPONDING TO 50% FROM THE CENTRE OF ROUND BARS.

For example, suppose that a steel possessing the hardenability characteristics shown in Fig. 61 were to be oil-quenched with a quenching severity of H = 0·35 ; then by reference to Fig. 63, it will be seen that maximum properties could be expected in bars, the diameters of which do not exceed $1\frac{1}{2}$ in.

It is sometimes advantageous to have a high surface-hardness and a relatively soft core, particularly in applications in which high surface loading occurs. This would indicate complete hardening to some fractional depth between surface and centre of the bar. For purposes of economy in selecting steel for such an application, it is helpful to be able to estimate whether a steel will harden to the required depth below the surface without undue hardening in the centre. This may be done through the use of the charts available, in which is shown the relation between Jominy depth and bar size that will harden to various fractional distances between centre and surface. For instance, assume service requirements are such as to require

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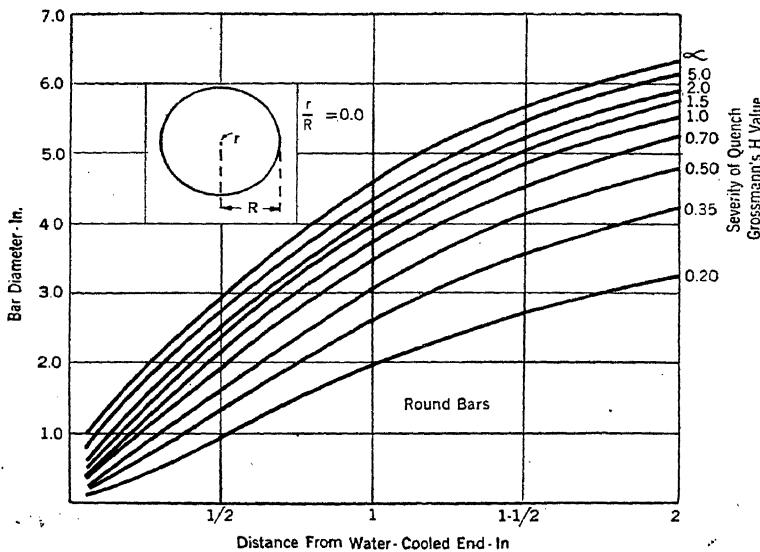
FIG. 63. LOCATION ON END-QUENCHED JOMINY HARDENABILITY SPECIMEN CORRESPONDING TO THE CENTRE OF ROUND BARS.
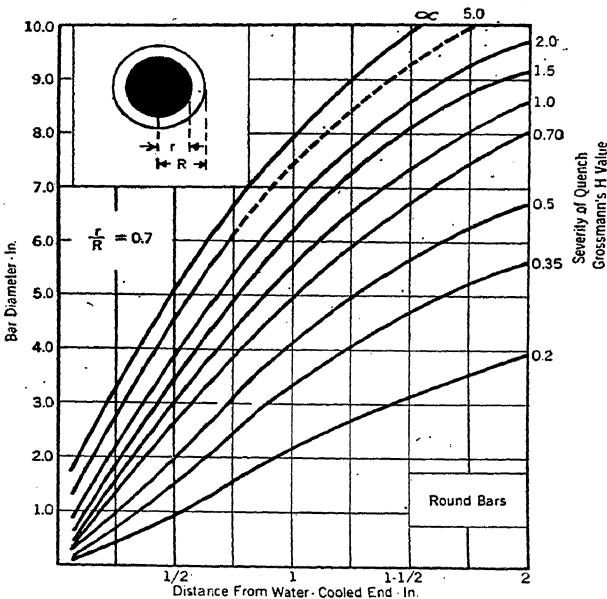
FIG. 64. LOCATION ON END-QUENCHED JOMINY HARDENABILITY SPECIMEN, CORRESPONDING TO THE CENTRE OF ROUND BARS.

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complete hardening to a depth of 0·6 in. below the surface of a 4 in. diameter bar. This would mean that the steel would have to harden fully at a fractional distance of $\dfrac{r}{R} = \dfrac{1\cdot4}{2\cdot0} = 0\cdot7$ from centre to surface. If the steel has a Jominy hardenability curve similar to that shown in Fig. 60, in which maximum hardness is maintained to a Jominy distance of $\frac{9}{16}$ in., then by reference to Fig. 64 for a fractional distance of $\dfrac{r}{R} = 0\cdot7$, it can be estimated that a severity of quench or H value of about 1·7 would be required. According to Table X, this rate of cooling could be obtained by agitating the bar while quenching in water.

Lamont also shows how the hardness to be expected in the core, at greater depths than 0·6 in. from the surface may be estimated and also how the behaviour of square bars, flat bars or plates may be predicted. Although the relations are not given with the degree of completeness shown in charts for round bars, an approximation sufficient to estimate roughly the extent of hardening may be gained from the diagrams provided.

STEEL SPECIFICATIONS

Greswolde van Dyke* has discussed the relative merits of basing steel specifications on analysis or performance.

The performance of a steel may be defined as including machineability, response to heat-treatment, specified physical properties and other characteristics.

Response to heat-treatment (" hardenability ") may be ascertained by means of the Jominy end-quench test, which is sufficiently reliable to indicate how a steel will perform in standard or commercial applications and therefore to serve as a yardstick whereby the performance of emergency and other new alloy steels may be checked.

The performance of a steel is, in the main, governed by composition but control of such performance by analysis is difficult.

The surface-hardness as developed by quenching is, in the main, governed by carbon content, while depth of penetration, if not exclusively, is largely governed by the alloy content, as well as by grain-size. There are other factors, not easily identified, which affect the hardenability of steel. Thus, two steels, of almost identically similar composition and grain-size, will have different hardenabilities. This individuality in a " heat " of steel, giving it greater hardenability than other " heats " of almost identically similar composition, is difficult to define and impossible to specify in terms of analysis or any other controllable factor.

Therefore, it would appear to be logical to specify steel on the basis of its hardenability, rather than indirectly by specifying analysis and grain-size.

The following example is given :—

The steel selected on past experience is A-3135, grain-size 5 to 8. From the same mill, the following two heats might be supplied, viz. :—

* *The Iron Age, Oct. 14, 1943, pp. 71-72.*

	Heat A. %	Heat] %
Carbon	0·34	0·38
Manganese	0·62	0·76
Silicon	0·22	- 0·34
Nickel	1·15	1·40
Chromium	0·57	0·74
Grain-size	No. 8	No. 5

TABLE XX.

Both these heats fall within the published analysis limit of A-3135. There would, clearly, be a considerable difference in hardenability of the two steels, especially in depth of hardness-penetration. Heat A might be too low and heat B too high in hardenability for the destined application.

Had these steels been ordered as Type A-3100, with a reasonable hardenability range specified, known to be suitable for the application, both heats could be applied and satisfactory results obtained without change in heat-treatment.

If a reasonable hardenability range were specified, and if the mill were not confined to a specific analysis range, it would probably have a better chance of furnishing a satisfactory product on the hardenability basis.
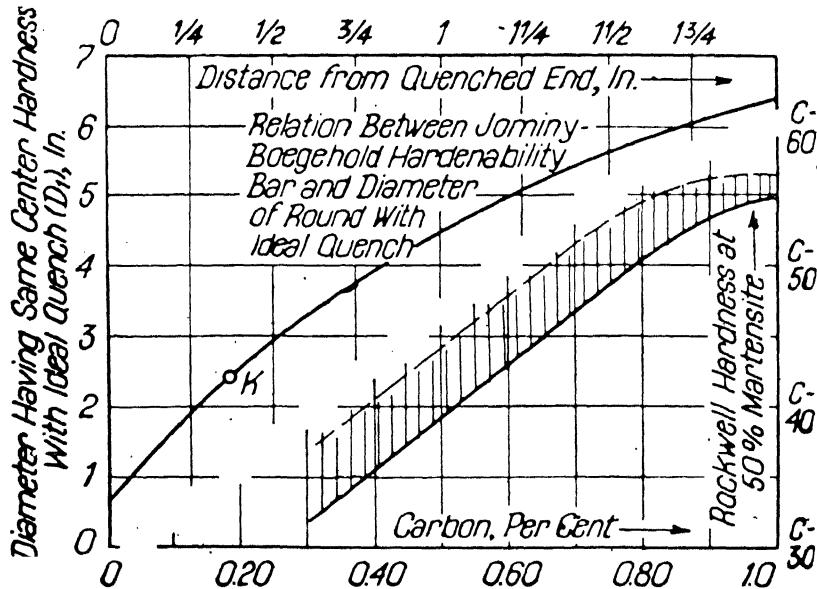
FIG. 65. (UPPER LEFT).—RELATION BETWEEN JOMINY HARDENA-BILITY BAR AND DIAMETER OF ROUND BAR WITH " IDEAL " QUENCH. (LOWER RIGHT.)—EFFECT OF CARBON PER CENT. ON ROCKWELL (C) HARDNESS AT 50 % MARTENSITE.

In Fig. 65 the lower diagram shows how carbon affects hardenability, the shaded area representing the range of hardness intensity for any given carbon content. The upper diagram shows the diameters of bars having a

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common hardness value at their centres, when ideally quenched, plotted against corresponding Jominy values obtained on end-quenched bars.

Hardenability of a steel may be designated by a code indicating the distance from the quenched end of the specimen within which the hardness is greater than is required for obtaining an acceptable quenched and tempered structure.

Examples.

(a) 0·35% *carbon steel.* Hardness above Rockwell C.50 for a distance of $1\frac{1}{2}$ in. (24/16 in.) from the quenched-end.
Index of hardenability J.50 = 24 (sixteenths).

(b) 0·15% *carbon steel.* Hardness above Rockwell C.30 for a distance of $\frac{5}{16}$ in. from the quenched-end.
Index of hardenability J.30 = 5 (sixteenths).

Welding Steels.

G. A. Timmons* proposes to use the Jominy test for a purpose just the reverse of its usual application, viz., for evaluating welding steels as regards their ability to be held *below* a specified maximum hardness value, when subjected to heating and cooling conditions similar to those obtaining during the welding operation. In that operation, the parent metal is required *not* to harden unduly when that part of it which is adjacent to the weld is subjected to varying cooling rates from a series of temperatures between the melting point of the steel and room temperature.

Chapter 4

HARDENABILITY

Austin, Van Note and Prater* have found that the effect of austenitic grain-size on hardenability, expressed in terms of critical cooling velocity, is essentially linear, whereas the relationship between the amount of an added third element and hardenability is not linear. Their experiments were made on hyper-eutectoid carbon steels of high purity. The increasing potency of third elements, added in amounts up to 5%, in their effects on critical cooling velocity, was found to be in the following order, viz.: nickel, copper, silicon, manganese, aluminium and chromium.

In Fig. 69 is shown the marked effect on hardenability attributed to very small additions of boron to a plain carbon steel.
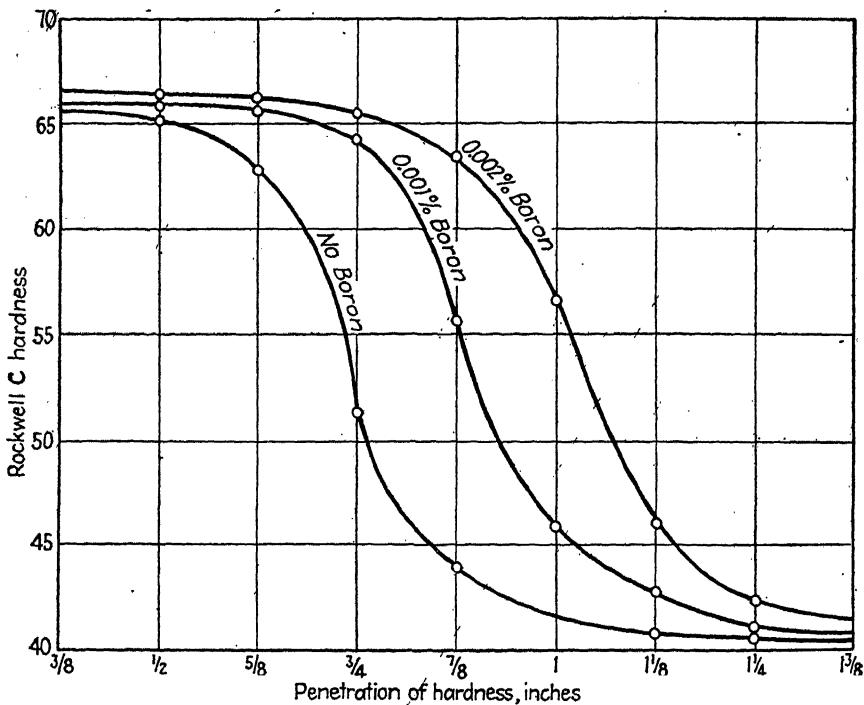
FIG. 66. EFFECT OF BORON ON HARDENABILITY CURVES OF 0·65% CARBON STEEL.

* Am. Soc. Met. 1942 (Oct.), pre-print No. 2.

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Grain-size, the McQuaid Ehn Test.

The purpose of this test is to ascertain for purposes of comparison, the grain-size of various steels, as affected by various conditions of heat-treatment.

In the heat-treated state, the grain-size of a steel often defies measurement, which, however, is rendered possible by this test. Whatever the austenitic grain-size of a steel may be, when the transformation to pearlite occurs, on cooling, the latter will have the same grain-size. If the steel be quenched at a suitable cooling rate, a network pattern may be obtained, as the result of nodular troostite forming at the crystal boundaries, enclosing grains of martensite. Such a structure lends itself to grain-size measurement. In a hypo-eutectoid steel containing excess ferrite or in a hyper-eutectoid steel containing free cementite, the separation of either of these phases, at the crystal boundaries, facilitates grain-size measurement in the structure obtained by slow-cooling.

The McQuaid Ehn test aims at securing a cementite network by carburizing the steel—the grain-size of which is to be determined—at a temperature at which grain-growth cannot occur. It is necessary to use a carburizing compound which will produce a sufficient amount of excess cementite to form a clearly recognisable network round the grains of pearlite, when viewed at a magnification of X100. The standard solid activated carburizing " compound " in commercial use is usually satisfactory for this purpose. There should be a sufficiently long period at 925° C. to give a case depth of 0·050 in. The rate of cooling can be determined only by experience. Plain carbon steels can be treated in the ordinary commercial procedure and cooled in the container. Air-hardening steels, such as the high nickel and chromium nickel-molybdenum steels, must be allowed to cool in the furnace. In laboratory tests, all test-pieces should cool in the furnace and, in the case of high alloy steels, a retarded rate of cooling of the furnace is necessary. The following minimum cooling times have been found satisfactory, viz. :-

Plain carbon steels	3 hours.
S.A.E. 2315 ($3\frac{1}{2}$% nickel)	4
S.A.E. 4615 (2% nickel, 0·9% chromium)	
S.A.E. 2512 (5% nickel)	8
Krupp steel (4% nickel, 1·5% chromium)	16

Grain Size and Hardenability.

Hardenability is affected by grain-size. It has already been pointed out that a fine-grained steel is liable to be imperfectly hardened, owing to the production of soft spots. By a fine-grained steel is meant one in which the crystal grains of austenite are small at the moment of quenching. Apart from this irregularity, such a steel is more satisfactory when heat-treated, as regards toughness, than a steel which, coarse-grained at the moment of quenching, has higher hardenability, but possesses less toughness, in its heat-treated state.

Grain-size is usually reported in numbers which represent the average number of grains per square inch at a linear magnification of 100.

No.—1	0	1	2	3	4	5	6	7	8	9	10	11	12
Grains per sq. in. .. $\frac{1}{4}$	$\frac{1}{2}$	1	2	4	8	16	32	64	128	256	512	1024	2048

Standard charts have been prepared for estimating grain-size (Am. Soc. Testing Materials, Standards (1933), p. 898). These charts can be used

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for different magnifications by correcting the measured grain-size numbers in the following way :—

Magnification	..	25	50	100	200	400	800
Correction No.	..	− 4	− 2	0	+ 2	+ 4	+ 6

A duplex structure (*e.g.* ferrite +pearlite) can be described by reporting both grain-sizes, and their approximate percentages of the total area.

Steel, like every other metal, is an aggregate of crystal-grains of varying shape, firmly cemented together by cohesive force. Each of these crystal-grains has an internal symmetry which renders it crystalline. In any piece of any steel the crystal-grains are never all of equal size. The size varies between a maximum and a minimum, and the smaller the difference between these two extreme sizes the more uniform is the structure. The grain-size, when regarded either as the size of the average grain, or the number of grains per unit volume, can vary between very wide limits.

When a steel is heated to the temperature at which, whatever its previous structure may have been, it transforms to austenite, it is still an aggregate of crystal-grains, which now, however, are (*a*) smaller than before the transformation, and (*b*) all of the same uniform chemical composition. At the temperature of formation little or no grain-growth occurs, *i.e.*, the initial grain-size undergoes no appreciable change. This initial grain-size is not always the same under all conditions. It is affected in the following ways :—

The presence of alumina (resulting from the use of aluminium as a deoxidizer in the production of the steel ingot) is regarded as an effective agent in inhibiting the growth of austenite grains.

Increase in the initial grain-size of austenite is caused by an increase in grain-size of the pre-existing constituents, *i.e.*, for low, medium or high carbon steels. This pre-existing grain-size will have been produced by heat-treatment or by finishing temperature in hot-working. In cold-worked steels, annealed below the temperature of formation of austenite, a grain-refinement has already occurred, so that if and when such steels are caused to undergo the transformation to austentite (the temperature of which is above the temperature of recrystallization brought about by annealing after cold-working), their small grain-size will result in the initial grain-size of the austenite formed from them, being small also.

By raising the temperature still higher, the crystal-grains grow by feeding on each other. This is called grain-growth or coarsening, and the temperature at which and the manner in which it occurs, are governed by several factors.

The coarsening temperature, viz., that at which grain-growth begins, varies between 870° C. and 1150° C. It is affected (*a*) by the pre-transformation grain-size, that is to say, a steel which is fine-grained before transformation to austenite will not coarsen at so low a temperature as a steel having a pre-existing coarse grain. This fineness of grain may have resulted from heat-treatment, from hot or cold working, from composition or from a combination of these factors.

Coarsening temperature is also affected (*b*) by the normalizing temperature. Standard normalizing, *i.e.* at temperatures not far above the critical temperature, will usually lower the temperature of coarsening in subsequent

heat-treatment, as compared with steel which has not been normalized at all or has been normalized at an unusually high temperature. A slower rate of cooling than that associated with normalizing, *e.g.* in the furnace after annealing, or a faster rate of cooling, *e.g.* in oil or water, usually causes a lowering of the coarsening temperature, as also does hot-working.

The mechanism of coarsening varies. It may be *uniform* in type or *exaggerated*. This difference is illustrated by Grossmann,* together with the following explanation :—
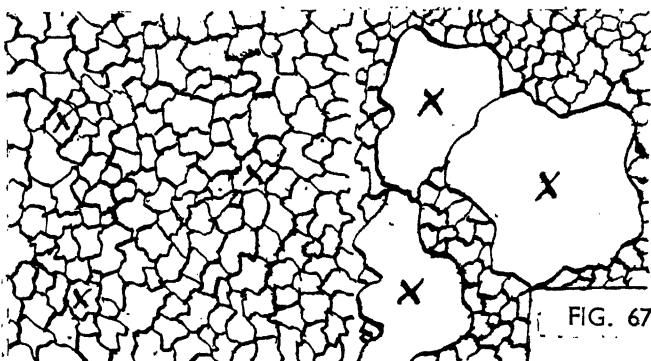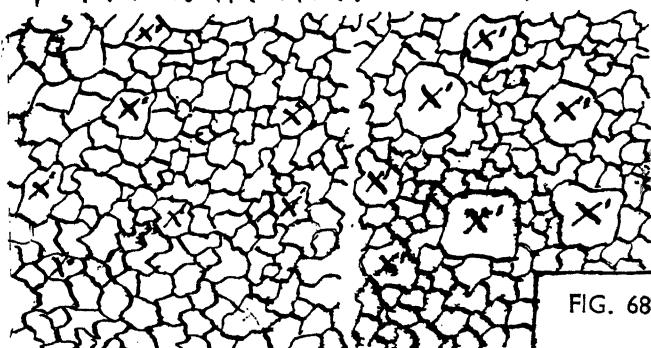
FIG. 67. EXAGGERATED " COARSENING. GROWTH PROCEEDS FROM A FEW CENTRES (MARKED " X ").

FIG. 68. " UNIFORM " COARSENING. GROWTH PROCEEDS FROM NUMEROUS CENTRES (MARKED " X^1 ").

Consider Fig. 67 (left half) to represent the initial austenite grain-size, before coarsening. Upon heating the austenite to a higher temperature, grain-growth may begin at perhaps three favoured positions, marked " X " in Fig. 67. The growth would continue with the gradual absorption of the surrounding grains until after a period of time there would be three large grains as in Fig. 67 (right half). On the other hand, grain-growth might begin at a very large number of favoured positions, such as those marked

* " *Principles of Heat-Treatment,*" *p.* 1

"x^1" in Fig. 68 (left half). After a period of time the appearance would be as in Fig. 68 (right half), showing a larger number of growth-grains, but the growth is not so great as in Fig. 67. *Uniform coarsening* may therefore be considered to be growth from a greater number of centres than in the case of *exaggerated coarsening*.

A. S. Jameson* provides evidence of the superior toughness of fine-grain over coarse-grain, instancing case-hardened steel (*see* Table XVII), and of the superiority of coarse-grain over fine-grain as regards hardenability of such steels (Fig. 69). Of the steels represented in Fig. 69, the coarse-grained specimens had grain-size numbers 2 and 4, and the fine-grained specimens had numbers 6 and 7.
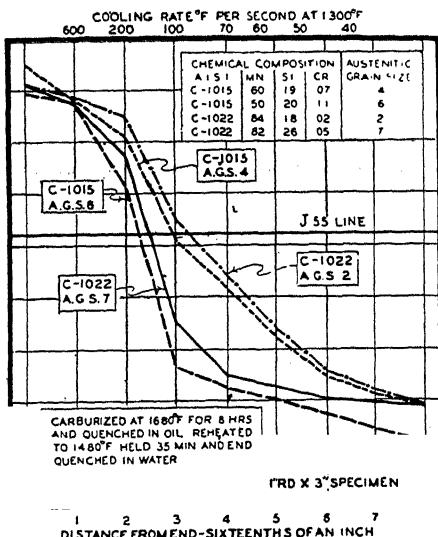
FIG. 69. EFFECT OF AUSTENITIC GRAIN-SIZE ON HARDENABILITY OF CARBURIZING STEELS. UPPER PAIR OF CURVES ARE FOR COARSE-GRAIN AND LOWER PAIR FOR FINE-GRAIN STEELS.

At 1300° F. (705° C.) a fine-grained steel requires a cooling rate about 100° F. (55° C.) per sec. faster than a coarse-grained steel (of the same chemical composition) to produce a part of satisfactory hardness, say Rockwell C. 60.

This does not necessarily make coarse-grained steels a preferable choice for carburizing purposes. Apart from their inferior toughness, already mentioned, they are more likely to have surface seams, which means higher costs for removal by chipping, when preparing blooms or billets for subsequent rolling operations. These disadvantages practically eliminate coarse-grained carbon steels from consideration as a substitute for alloy steels. On the other hand, fine-grained steels are so inferior in hardenability that they, too, are ineligible as substitutes, except for parts of very small section, say, under $\frac{1}{4}$ in. in thickness.

* " *The Iron Age.*" *1943, 151, pp. 47–56.*

TABLE XVII.

Effect of Grain-size on Toughness of Case-hardened Mild Steel of Composition, Carbon 0·18%, Silicon 0·20%, Manganese 0·5%.

Treatment, after carburizing at 927° C.	Coarse-grained A.S.T.M. No. 2.				Fine-grained A.S.T.M. No. 7.			
	Grain-size No. (by fracture).	Hardness Rockwell C. (average).	Case Depth	Impact Test. Energy Absorbed Ft.—lbs.	Grain-size No. (by fracture).	Hardness Rockwell C. (average).	Case Depth	Impact Test. Energy Absorbed Ft.—lbs.
Oil-quenched from box. Re-heated to 777° C. and brine-quenched. Tempered at 162° C.	2	64	In. ·035	7	7	64	In. ·030	17
Slowly cooled in box. Re-heated to 915° C. and oil - quenched. Re-heated to 777° C. and brine-quenched. Tempered at 162° C.	2	64	·043	6	7	64	·038	13

The problem may be met by using steels of higher manganese content, *e.g.* A.I.S.I. C 1022 with a range of 0·7–1·0% manganese would be selected in preference to A.I.S.I. C 1015 with a range of 0·30–0·50% manganese.

Fig. 70 shows the effect of increasing manganese content on the hardenability of carburized steel, as determined by the Jominy end-quench test. The curve gives the depths of hardening (to a hardness of 55 Rockwell C. or over) for varying manganese contents in a low carbon steel.

COOLING RATE °F PER SECOND AT 1300°F

DISTANCE FROM END(SIXTEENTHS OF INCH)

LOW CARBON-OPEN HEARTH-FINE GRAIN
LESS THAN .050 S   .14 -.19 CR.
LESS THAN .040 P   .15 -.25 SI.
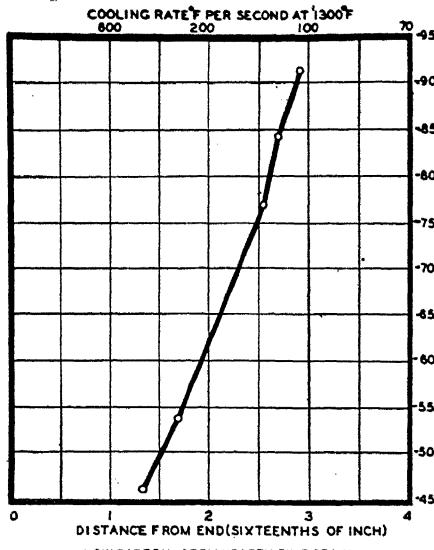
FIG. 70.   EFFECT OF MANGANESE CONTENT ON DEPTH OF HARDENING IN HARDENABILITY TESTS.

The question of grain-size is important, also, in the case of medium carbon steels.

Increase in grain-size, accompanied by increase in hardenability, permits the use of larger sections when hardened—an important consideration with oil-quenched steels. Fine-grained steels are preferable if they have to be water-quenched, as such steels distort less and are not so prone to crack when hardened.

With steels of a given carbon content there is a limit to the size of section in which a given hardenability can be obtained. By increasing the carbon content, this limit can be exceeded as shown in Table XVIII, the data on which were obtained on fine-grained steels. The size of section hardenable to the same extent is still greater when coarse-grained steels are used (see Table XIX).

The hardness number of 45 Rockwell C. was selected, as it was assumed to be the lowest hardness at which, after quenching, the structure would be at least 50% martensitic. In this condition, tempering can produce a uniform sorbitic structure, such as would be found associated with steels in the specified hardness range 22 to 38 Rockwell C.

Data justifying the choice of this hardness value are obtained from Fig. 71, in which hardness penetration curves are given for steel 3140, showing that test-pieces machined from the centre of a 3-in. diameter (9 in. long) tempered bar possessed satisfactory tensile properties, even though the Rockwell C. hardness, as quenched, averaged only 43. Test-pieces taken nearer to the surface, in which the hardness as quenched did average 47, gave even better results.

TABLE XVIII.
*Influence of Carbon Content on Hardenability
in terms of Size of Bar.*

Carbon, per cent.	Maximum Diameter of Bar, which will harden at surface to at least 45 Rockwell C. Inches.		Diameter of Bar which will give maximum Tensile Properties. Inches.	
	Oil.	Water.	Oil.	Water.
0·35		$2\frac{1}{4}$		$\frac{3}{16}$
0·41		$3\frac{1}{2}$		$\frac{3}{8}$
0·45	$\frac{7}{16}$	$3\frac{3}{4}$	$\frac{1}{8}$	$\frac{1}{2}$
0·52	$\frac{5}{8}$	$3\frac{13}{16}$	$\frac{3}{32}$	$\frac{1}{4}$
0·58	$\frac{15}{16}$	$3\frac{7}{8}$	$\frac{1}{4}$	$\frac{3}{8}$

Quenching temperature 843° C.   Steels fine-grained.

TABLE XIX.
Influence of Grain-size on Hardenability of a 0·45% Carbon Steel.

	Fine-grained. Austenitic Grain. Size No. 7.	Coarse-grained. Austenitic Grain. Size No. 2.
	In.	Ins.
Maximum diameter of bar which will harden at surface to at least 45 Rockwell C. when oil-quenched from 815° C.	$\frac{7}{16}$	$1\frac{1}{4}$
Diameter of bar which will give maximum tensile properties when oil-quenched from 815° C. ..	$\frac{1}{8}$	$\frac{7}{16}$

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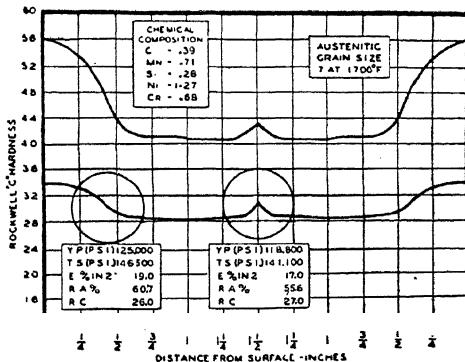
FIG. 71. HARDNESS-DISTRIBUTION CURVES OF 3-IN. DIAMETER STEEL
BAR, AS QUENCHED FROM 843° C. (TOP CURVE) AND AFTER TEMPERING
AT 482° C. (LOWER CURVE). CIRCLES INDICATE POSITIONS AT WHICH
TENSILE TEST SPECIMENS WERE SELECTED.

Hardenability is governed by the following factors :—

(a) Composition.
(b) Grain-size.
(c) Quenching severity.

(a) *Composition.*

Carbon content is fundamentally important. Nature and percentage of alloying element or elements and combinations thereof, are supplementary factors.

(b) *Grain-size.*

This, in turn, is often governed by composition.

(c) *Quenching Severity.*

This governs the rate of cooling, which is affected by hardening temperature, degree of uniformity of temperature, conditions of surface, surface area per unit volume (size of section) and mass of piece. The nature of the quenching medium, its temperature and rate of movement over the surface of the piece, are also of great importance.

Grossmann and Stevenson* examined steels of a wide range of compositions and therefore of hardenabilities in order to elucidate the effect of grain-size. Each steel was made to have different grain-sizes by heating to a series of temperatures prior to quenching, the hardenability being subsequently recorded. It was found that the greater the hardenability —as caused by chemical composition—the more was it affected by a change in grain-size.

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THE S-TRANSFORMATION CURVE

The temperatures at which the critical thermal changes occur in iron-carbon alloys, as shown in the iron-carbon equilibrium diagram are not independent of the rates of heating and cooling. They are raised by fast rates of heating and lowered by fast rates of cooling. With a rate of cooling which, whilst not conforming to the *very* slow rate necessary for the establishment of equilibrium, may yet be described as a slow rate, the amount of "undercooling" or "super-cooling" which occurs, is small. That is to say, the Ar_1, Ar_2 and Ar_3 points are depressed only slightly. The Ar_1 (eutectoid inversion) point is the one at which austenite transforms to pearlite. The slower the rate of cooling the more coarsely lamellar is the pearlite and the lower its tensile strength and hardness. As the rate of cooling and the amount of depression of the Ar_1 point increase, the pearlite becomes progressively finer, stronger and harder. In a hypo-eutectoid steel the amount of pro-eutectoid ferrite, or, in a hyper-eutectoid steel, the amount of pro-eutectoid cementite, diminishes and may even fail to precipitate—as separate phases—altogether, at very fast rates of cooling, in alloys in which the carbon content is not greatly different from that of the eutectoid alloy.
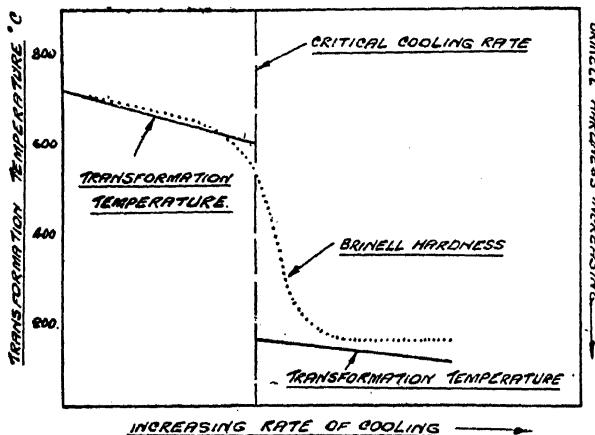
FIG. 72. EFFECT OF RATE OF COOLING ON AR′ (LEFT) AND AR″
(RIGHT) TRANSFORMATION TEMPERATURES.

As the degree of "undercooling" increases with progressively increasing rates of cooling—the transformation may be lowered to temperatures below 540° C.—the celerity with which it then occurs also increases, the opportunity for the development of coarse lamellæ at these lower temperatures is restricted and the resultant structure may be so fine as to be incapable of being resolved except under the highest magnifying powers. This is usually described as *sorbitic pearlite*.

The rates of cooling attained in hardening operations, *e.g.* water-quenching from temperatures above the critical point, result in a depression of the transformation point (by "undercooling") to temperatures approaching room temperature, the product being, of course, martensite.

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The progressive influence of increase in rate of cooling on the temperature of the austenite → pearlite transformation is shown in the upper (left-hand) transformation curve in Fig. 72. When, as the result of increasing the rate of cooling, a single transformation occurs, due to " undercooling ", it is designated Ar' and represents the appearance of a new transformation product, the constituent *troostite*, called by American metallurgists " fine pearlite ". As has already been mentioned in Chapter I, the nature and properties of this constituent vary with the temperature of its formation. At the higher temperatures it has a nodular, dark etching form ; at lower temperatures, an acicular, dark etching form. At still lower temperatures, it has a feathery appearance and in this form the name " Bainite " has been suggested.

A rate of cooling, in which the Ar' transformation is allowed to occur is slower than the *critical cooling rate*. The latter has been defined as the slowest rate which will give a fully-hardened martensitic structure, *i.e.* a structure which, being free from troostite, has not experienced the Ar' change during its formation. The heat-evolution associated with the formation of martensite is indeed at a much lower temperature than that of Ar' ; it is at or below 150° C., and is designated Ar". The lower transformation curve at the right of Fig. 72 is the Ar" curve and shows that this change also is progressively depressed by increase in rate of cooling.

The dotted curve (in Fig. 72) shows the disparity in hardness between the transformation product (troostite) resulting from the Ar' change and that resulting from the Ar" change (martensite).

The transition from the high-temperature product to the low temperature product, although sharp, is continuous. This is due to the fact that over a certain narrow range of cooling rate, both Ar' and Ar" changes can occur in the same piece of steel, resulting in a mixture of the two products, the softer troostite and the harder martensite. The hardness will increase rapidly with the proportion of martensite present. In a very small piece of hardened carbon steel, in which the rate of cooling has equalled or exceeded the critical cooling rate, the structure will be all martensite, and the transformation temperature Ar" will fall on the lower curve of Fig. 72. In a larger piece of the same steel, the outer shell will conform to the same conditions as the small piece, but the inner core will have cooled at a slower rate than the critical cooling rate and some of the austenite will have undergone the Ar' transformation of the upper curve, resulting in the formation of troostite.

Quantitative data provided by E. C. Bain*, as plotted in Fig. 73, show two outstanding features, viz. :—

(a) Progressive lowering of the transformation temperature as rate of cooling increases ;

(b) Progressive diminution in the time intervals preceding transformation, *i.e.* in the pre-transformation times.

(a) In the case of a steel of eutectoid composition, the extent of the depression, as cooling rates increase, is as follows :—

Cooling rate.	Depression ° C.
$\frac{1}{8}$° C. per min.	8
$\frac{1}{2}$° C ,,	11
2° C. ,,	16

' *Metals and Alloys.*" 1937, 8, 22.

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The time scale in Fig. **73** is logarithmic in order to show on the same chart the effects of very fast and very slow rates. This accounts for the curvature in what are actually uniform rate cooling-curves (D, C and B). The microstructures corresponding to these rates of cooling are as follows :—

Rate	Microstructure	Brinell Hardness
D	Coarse pearlite	210
C	Sorbitic pearlite	315
B	Troostite ..	429
A	Martensite ..	circa 700

At a rate faster than that shown by curve B, (Fig. 73) but still not quite so fast as the critical cooling rate, the microstructure is predominantly martensitic with a small amount of troostite and a Brinell hardness number of about 670. In this instance, only a small amount of austenite has transformed to troostite at the higher (Ar') temperature, the larger proportion having transformed to martensite at the lower (Ar") temperature. At the still faster cooling rate shown by curve A, none of the austenite has transformed to troostite, only the Ar" arrest having been experienced (just under 100° C.) at which all the austenite transformed to martensite. In carbon steels, the Ar" point cannot be depressed to below room temperature, however great the quenching severity, therefore little or no retained austenite is to be found in such steels, hardened under commercial conditions. The addition of alloying elements, however, makes it possible for austenite to be retained in the structure of a hardened steel.

The experimental work of Bain and his collaborators has formed the basis for the establishment of the time-temperature-transformation curves, known from their shape as S curves. They give, in fuller detail than Fig. 73, the times taken by austenite to begin to transform, and also the times taken to complete transformation, at all temperatures above room-temperature (*see* Fig. 23, Chapter I). The data are obtained by the use of heated quenching baths, held at constant temperature. Steel specimens are heated **to** and held at a temperature above the equilibrium transformation temperature, until they have become a uniform solid solution of carbon in gamma iron or, to use the American term, until they have become "austenitized". When quenched in a heated bath kept at constant temperature, below that of the equilibrium transformation temperature (sometimes called "subcritical") a certain interval of time elapses before transformation begins. If, in specimens taken from the hot quenching bath and water-quenched, the resulting structure is all martensite, it is clear that the Ar' transformation has not yet begun. When it has begun, the water-quenched specimen will show a small amount of troostite in addition to martensite. Other specimens, still in the hot quenching bath, continue to undergo transformation, as can be shown by the amount of troostite shown to be present when the specimens are water-quenched after increasing intervals of time. When all of the austenite has undergone transformation, a water-quenched specimen will show 100% troostite, *i.e.* the absence of martensite from the structure indicates that no austenite remained in the specimen when it was taken from the hot bath prior to water-quenching. S curves show that the pre-transformation times diminish at first, as the temperature of the hot quenching
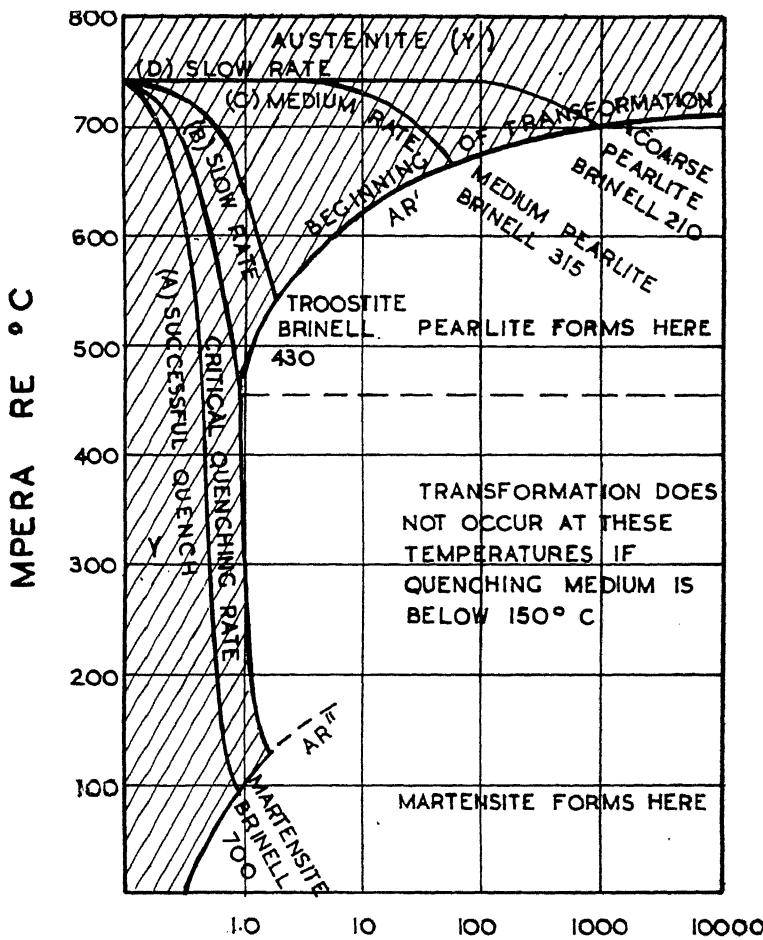
FIG. 73. CHART SHOWING TIME AND TEMPERATURE OF INITIAL
TRANSFORMATION OF A EUTECTOID STEEL. COOLED AT VARIOUS
UNIFORM RATES. (*E. C. Bain.*)*

bath is lowered. Between 600 and 500° C. the time is very short indeed, as shown by the proximity of the bulge in the curve to the temperature axis, this representing the most critical stage in the hardening of carbon steels, especially those of hypo-eutectoid constitution. If a sufficient speed of cooling can be reached, so as to by-pass this bulge, greater latitude in cooling rate is thereafter allowable, as will be appreciated from a glance at the increase in pre-transformation times as the temperature decreases below the bulge. At 300° C., for instance, specimens do not begin to transform

* "*Metals and Alloys.*" *1937, 8, 22.*
For Curve B in Chart, read "Quickened Rate."

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until after the lapse of about one minute. Hereabouts, the time required for *completion* of the transformation is increasing, occupying several days at a temperature just below 200° C.

The higher the carbon content the further to the right is the Ar' bulge of the S curve, thus accounting for the greater ease with which high-carbon steels can be fully hardened, as compared with medium-carbon steels.

At still lower temperatures, the direction of the curve is reversed, coming back again towards the temperature-axis. A steel which, by virtue of a

0.35% CARBON STEEL

C-035 Mn-0.37

AUSTENITIZED AT 1550 °F, GRAIN SIZE; 75 % 2-3 + 25 % 7-8
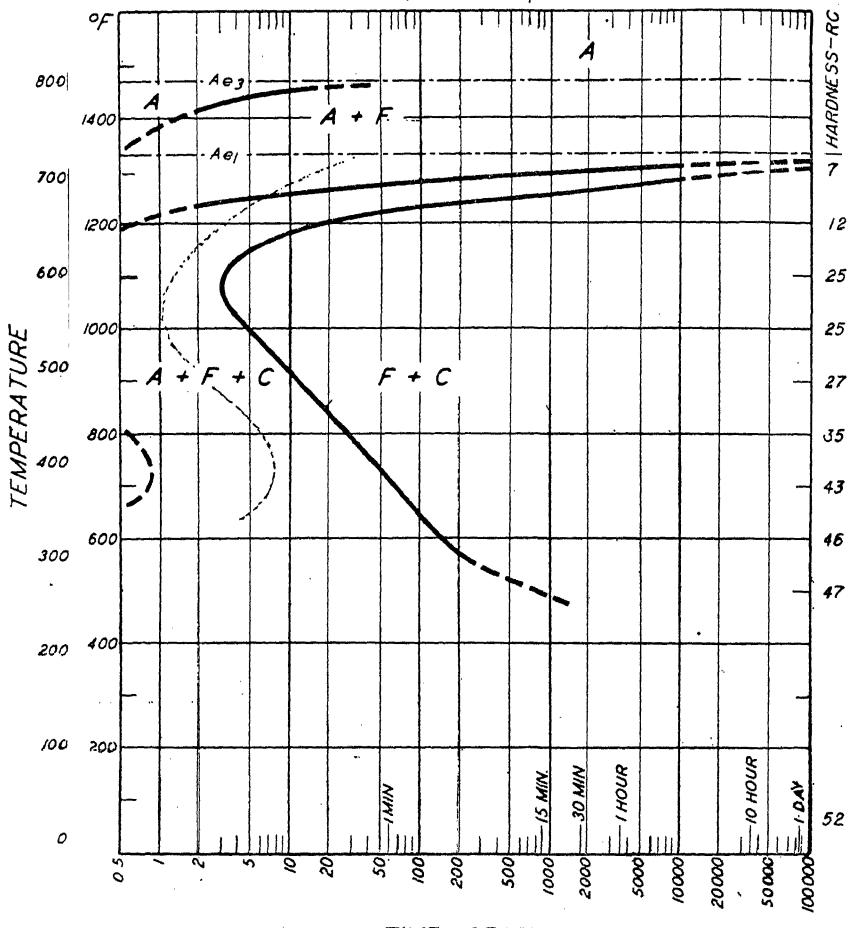
FIG. 74. S-CURVE OF 0·35% C STEEL.

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severe quench, has remained austenitic at a temperature nearly as far down as this reversed portion of the curve, will transform direct to martensite when it is allowed to reach the temperature (Ar″) of this part of the curve. This is what happens when the quenching severity is great enough to harden a steel fully.

By speeding past the left of the upper bulge of the curve, the Ar′ arrest is eluded, but the Ar″ arrest is inescapable.

If the initial cooling rate is a little too slow, the steel will be caught in
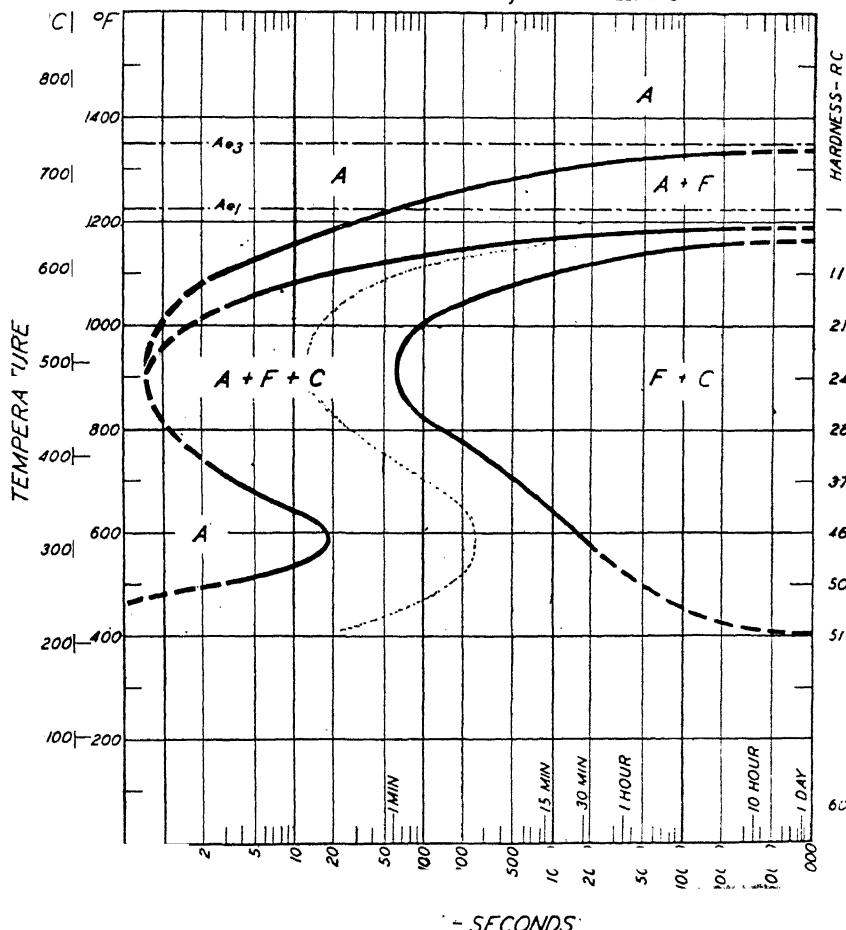
FIG. 75. S-CURVE OF ALLOY STEEL (

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the upper bulge, thus being compelled to surrender some portion of its austenite to form troostite; the remainder transforms to martensite when the Ar" curve is reached.

In the plain (0·35% C.) carbon steel curve (Fig. 74) it is clear that no opportunity is given to catch the steel in the austenitic condition until quenching baths at temperatures lower than the Ar' transformation temperature are employed and then only in a very short temperature range, round about 400° C. (the maximum pre-transformation time in this range being

FIG. 76. S-CURVE OF ALLOY STEEL S.A.E.4340.

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less than one second). In this diagram, as in Figs. 75 and 76, A = austenite, F = ferrite, C = cementite.

F + C represents any transformation product (*e.g.* pearlite, troostite, etc.) other than martensite.

In steel S.A.E. 2340 (Fig. 75) the presence of nickel is responsible for an increase in pre-transformation time intervals, so that it is possible by employing a quenching liquid at 300° C. to keep the steel in the austenitic condition for nearly 20 seconds.

In steel S.A.E. 4340 (Fig. 76), although less nickel is present, the carbon and manganese are higher and chromium and molybdenum are present in small amounts. The result is that the Ar' bulge has been pushed well over to the right, making it possible to keep the steel in the austenitic condition for nearly ten hours at about 540° C. Nickel, which extends the range of the gamma phase, as shown on the equilibrium diagram, has relatively small influence upon retarding the austenite $\rightarrow$ pearlite transformation. Chromium and molybdenum have much more powerful effects in this respect and, in addition, it is well established that combinations of such elements, which, in the aggregate, amount to no more than any one present singly, will have a more powerful influence than the latter.

Blanchard, Parke and Herzig* state that " of the many factors which influence rate of austenite decomposition at constant sub-critical temperatures, the most noteworthy are :—(*a*) austenite grain-size, (*b*) rate of formation of pearlite nuclei, (*c*) rate of growth of pearlite grains, (*d*) undissolved carbides and non-metallic inclusions, (*e*) degree of heterogeneity (*f*) chemical composition. The last-mentioned exerts the most important effect on the time for initiation and rate of transformation in the steel.

Arrested Quenching.

B. F. Shepherd† points out that the impression that carbon is only responsible for the maximum hardness that can be obtained in any steel and that hardenability from the standpoint of depth of hardening is a function solely of alloy content (in alloy steels) is incorrect.

The Jominy hardenability curves are flattened out for increasing carbon content for any series of steels in which the content of special alloying elements is constant.

Thus, for such steels, increase in hardenability can be obtained by increases in carbon content, without danger of producing hardening cracks or objectionable residual hardening stresses, if the following technique‡ be adopted :—

(*a*) An arrested (timed) quench in water.

(*b*) Transference to a molten salt bath, in which a timed immersion is given.

(*c*) Cooling in air to atmospheric temperature.

(*d*) Tempering in the normal way.

Trans. Amer. Soc. Met., 1941 29, 317-335.
† *"Iron Age." 1943, Jan. 28, p. 50, Feb. 4, p. 45.*
‡ *Named by Shepherd " Martempering."*

Steels of suitable composition may be quenched directly in molten salt,

By adopting such variations from conventional hardening technique, the failure of the latter to develop the maximum physical properties in the centre of heavy sections may be overcome. The same heating procedure (to produce homogeneous austenite), followed by the same cooling rate, will produce the same structure and hardness in steel of similar grain size, cleanliness, etc., no matter how the cooling rate was produced or in what ction of the piece it is located.

Similar structures mean similar hardnesses, but similar hardnesses do necessarily mean similar structures.

In the heating operation, to produce 100% austenite in hypo-eutectoid steels, the excess ferrite must be taken entirely into solution. In hyper-eutectoid steels, it may not always be desirable to cause the excess cementite to dissolve, but the pearlite must all transform to austenite.

The rate at which this occurs is dependent on chemical c In alloy steels the diffusion rate may be very low or the carbi to dissolve.   ıg tim   · ·   ·
bility, varies wit   ion.
          the   allowing the temperature to
                      critical cooling velocity and ·
physical properties.   overheating is less dangerous, grained steels—than heating to too *low* a temperature. Not only do
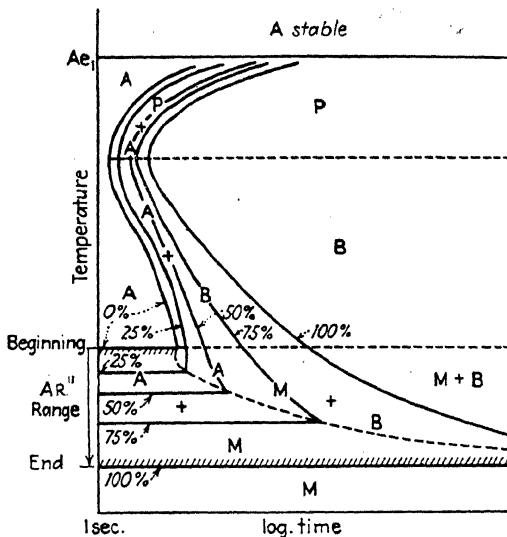
FIG 77. MODIFIED S-CURVE DIAGRAM. FULL LINES REPRESENT PERCENTAGE OF AUSTENITE TRANSFORMED.
A=AUSTENITE. P=PEARLITE (OR TROOSTITE).
B=BAINITE. M=MARTENSITE (OR TEMPERED MARTENSITE).
Ae₁=EQUILIBRIUM TRANSFORMATION TEMPERATURE AUSTENITE

media offer a wide range in cooling rates but, also, each liquid itself provides an appreciable range, depending upon its temperature or whether the piece is merely immersed, agitated, quenched in a spray or in a fixture under pressure. Oils vary considerably in type and quenching speed. Many articles have shapes so unsymmetrical that spray-quenching cannot be applied for fear of excessive distortion. Whatever quenching medium and method of application are chosen, the critical cooling rate of the parts to be hardened must be exceeded, though not greatly.

Excessively fast cooling rates cause internal stresses, high enough seriously to affect service life and to produce distortion, or even cracking. As a general rule the deeper-hardening the steel the slower the quenching rate of the recommended medium. Distortion will increase progressively with increase in quenching severity. Even though cracking does not occur,
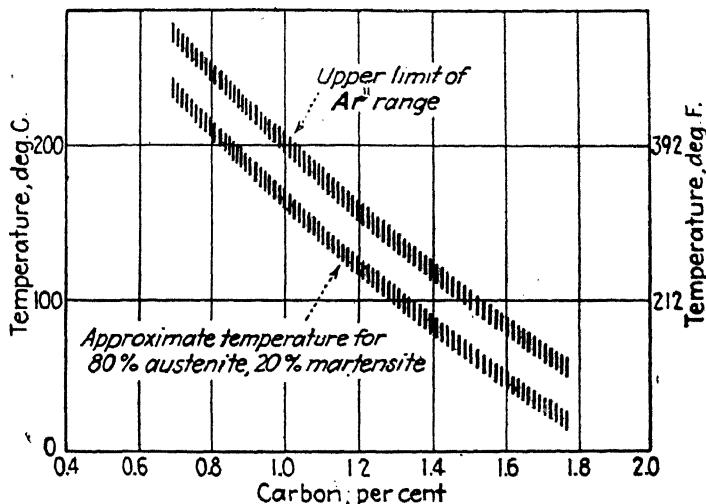
FIG. 78. INFLUENCE OF CARBON ON TRANSFORMATION TEMPERATURE (AUSTENITE→MARTENSITE) FOR PLAIN CARBON STEELS.

residual stresses are often present. If unrelieved by tempering they may be responsible for many service failures, even in parts of relatively simple shape. They are due, as previously explained, to volume-changes.* The differential rates of normal contraction caused by differential rates of cooling are aggravated by the expansion due to the formation of martensite. This constituent is so hard and unyielding that it is incapable of relieving stress locally, thereby spreading the load and lowering the unit stress and unit strain.

Only when martensite is accompanied by austenite are local yielding and consequent mitigation of the internal stress possible.

The formation of martensite occurs, not isothermally, but over a range of temperature, as shown in the modified S-transformation curve† (Fig. 77).

Isothermal transformation of austenite to intermediate constituents

* See Chap. 2, p. 50.
† Morris Cohen. A.S.M. 1940, 28, p. 565.

such as troosite, bainite, etc., is effected by rapidly cooling to some temperature within the transformation range and holding thereat long enough to allow the transformation to be complete. The temperature (Ms.) at which the formation of martensite is initiated by nuclei, varies within a narrow range, depending on its pre-hardening thermal history, grain-size, etc. The influence of carbon-content on the temperature of this range is shown in Fig. 78 (Greninger and Troiano).*

In Fig. 79, the temperature at which martensite starts to form (Ms) and the temperature at which its formation is complete (Mf) are shown in relation to various rates of cooling. As previously pointed out, the formation of martensite is attended by an expansion. When only a part of the steel is undergoing this expansion, whilst another part, not having reached the Ms point, is contracting, internal stresses are set up. These stresses will be greater the greater the temperature difference between the martensite already formed and the austenite in which no transformation has yet begun. If this difference be lessened or, better still, if the whole piece can transform to martensite and undergo expansion nearly isothermally, internal stresses will be obviated and cracking averted. There is a big advantage in causing martensite to form nuclei, well distributed in a cushion of ductile austenite, rather than allowing it to form a hard, rigid, unyielding outer shell, the expansion of which is incompatible with the contraction of the inner core. The greater the quenching severity (e.g. water) the greater the temperature gradient in the piece, and the greater the disparity between the completion of the martensite formation in the outer shell and its commencement in the inner core.

This is graphically shown in Fig. 79, which also illustrates the gentler gradient when oil is used as the quenching medium. Provided that the cooling rate exceeds the critical at the centre, even though it can never equal the cooling rate of the outer shell, martensite will be formed at the centre as well as in the outer shell and the piece will be hardened throughout. This can happen in the case of steel, even of thick section, and even when air-cooled, if of suitable composition. An air-hardened steel is more favourably placed as regards freedom from internal stresses, owing to the temperature difference, between the outside and inside, being so small that the whole mass passes through the transformation to martensite at practically the same temperature (Fig. 79).

If, however, composition and size of section be such that ordinary water-quenching would result in cracking whilst oil-quenching would leave the centre incompletely hardened, a technique may be devised to secure full hardening throughout, without cracking. This is effected by exceeding the critical cooling rate for the whole mass inside as well as outside, thus ensuring that the martensite formation shall begin, even at the centre, *but not until the temperature of the piece is uniformly low*. The final stage of cooling may then proceed, with a very much lower temperature gradient in the piece, so that no possibility remains of the metal, even at the centre, being able to escape conversion to martensite. This equalization of temperature of inside and outside of the piece by bringing it down rapidly to just above the Ms temperature, is effected by an arrested water-quench, *i.e.*,

Trans. Amer. Soc. Met., 1940, 28, p. 537. See also discussion by Morris Cohen (ibid) and Trans. Amer. Soc. Met., 1942, 30, p. 1,139.

taking it from the bath before its temperature has fallen to that of t water. Removal of the piece to a liquid bath (*e.g.* molten salts), which just above Ms temperature, allows temperature equalization to occur l heat-transference from centre to outside. So soon as that condition achieved, air-cooling can follow, the transformation to martensite beginnii at the outside at a temperature little different from that at which it finish in the centre.

Expansion of martensite grains as they form and grow can thus safely cushioned in the austenite matrix until the latter itself, in its tra formation to martensite, disappears.
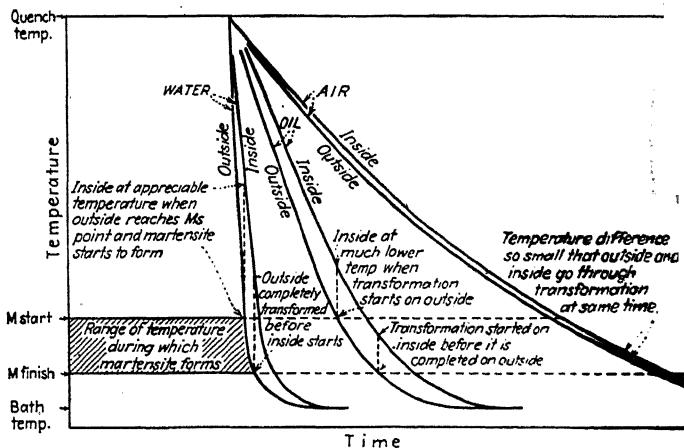
FIG. 79. COOLING RATES OF SURFACE AND CENTRE OF SIMILAR ROUND BARS OF STEEL, QUENCHED IN WATER, OIL AND AIR.

The timing of the arrested water-quench is obviously important. If too long, the outside layer transforms completely to martensite and the whole object of the method is lost. Care must be observed, when transferring wet pieces of steel from the water to the hot salt bath at or about 200° C. The sudden generation of steam results in the explosive scattering of the molten salt, which can easily cause dangerous burns.

These troubles may be avoided with some steels, the composition of which allows of direct quenching, from the hardening temperature, in molten salt (e.g. NE 8949).

Per cent.

Carbon	Manganese	Silicon	Chromium	Molybdenum
0·49	1·01	0·20	0·54	0·38

The salt must be at the appropriate temperature for promoting a cooling rate faster than that of the critical rate and yet permitting a leisurely equalization of temperature, the temperature of the bath being just above the Ms point. Thicker bars require longer times for equalization, *e.g.* at 204° C., 1-in. round bars require five minutes, 2-in. round bars eight minutes and 3-in. round bars thirteen and a half minutes.

Considerable latitude is possible as regards time of immersion before

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removing the steel for final air-cooling. This last stage of cooling in air need not always be practised. At 65° to 93° C. there may still be an appreciable amount of austenite left, and when reheated from this temperature range to the tempering temperature, sufficient time can be allowed for residual austenite to transform isothermally without being productive of internal stress.

The mechanical properties of a steel, treated in this way, are compared (in Table XI) with the properties of a piece of steel (adjacent thereto, when cut from the same bar), quenched in oil.

<div align="center">TABLE XI.</div>

Treatment.	Yield Point. tons/in.²	Tensile Strength tons/in.²	Elonga-tion. Per cent. in 2 in.	Reduction of area. Per cent.	Brinell No.	Izod Test. Ft. lbs.
Quenched at 815° C. in No. 2 soluble oil at 27° C. Tempered at 593° C.	65	70	19	57·3	321	48·5
Quenched at 815° C. in molten salt at 204° C. Held 5 mins. Cooled in air. Tempered at 593° C.	67	72·8	18	56·4	321	49·0

The quenching severity in oil is much less than in molten salt at the same temperature, as shown by the hardness-value, in the curves (Fig. 80).

Fifty per cent. of a quantity of ¾-in. round bars of steel (NE 8442) when heat-treated to give the mechanical properties shown in Table XI, gave trouble by cracking longitudinally, after oil-quenching. No cracking occurred when the molten salt quench was adopted, although the same hardness was obtained (Table XI).

The composition of the steel was as follows :—

<div align="center">Per cent.</div>

Carbon	Manganese	Silicon	Chromium	Molybdenum	Nickel
0·40	1·43	0·22	0·29	0·32	0·23

It is claimed for the process that

(a) Excellent depth of hardening and mechanical properties can be secured ; and

(b) By avoiding the cracking and distortion caused by the severe quenching of conventional hardening technique, the high hardenability obtained by increasing the carbon content of low-alloy steels, may be safely secured.

<div align="center">128</div>
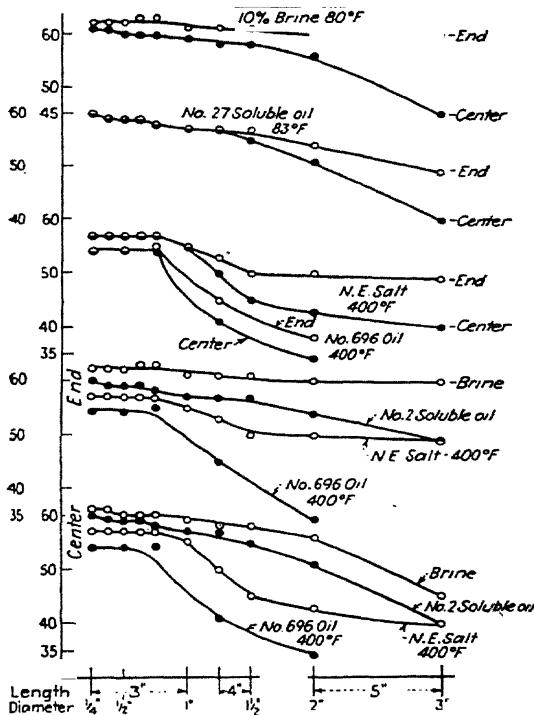
FIG. 80. HARDNESSES (ROCKWELL C) OF CROSS-SECTIONS OF VARIOUS SIZED ROUND BARS OF STEEL (NE 8744) QUENCHED FROM 815° C. COMPOSITION OF STEEL : C 0·43 %, MN 0·87 %, CR 0·50 %, NI 0·46 %, MO 0·27 %, P 0·03, % S0·008 %.

Chapter 5

MECHANICAL PROPERTIES (TESTING)

The behaviour of metals in structural and constructional applications, depends upon their mechanical properties. When subjected to external loads or forces, a metal article undergoes a change of dimensions, the magnitude and performance of which are governed by the magnitude of the load or force applied. The strains produced and the stresses creating them are measurable by testing and the mechanical behaviour of a metal is dependent on the inter-relationship between them.

No part in a metal structure or machine should suffer, in service, any permanent dimensional change. Such change results from a plastic condition of the metal and invalidates the calculations which the designer bases on the assumption that all anticipated stresses will be accommodated by elastic, *i.e.* temporary, dimensional change. Such change is necessarily very small indeed (usually less than 0·10%).

In service, however, perfect elasticity, although desired and often counted on, is somewhat illusory. Where complete recovery of the original dimensions, on releasing the load, is expected, a small permanent dimensional change often remains, decreasing slowly over an extended period of time. This is known as the " elastic after-effect."

A test-piece made from a single crystal of a metal is far weaker than a test-piece made from a polycrystalline metal, *i.e.* from an aggregate of smaller crystal grains. A single perfect crystal under stress yields internally thereto by a process of gliding along its slip planes, with little or no resistance. Sir Lawrence Bragg, in making reference to the experiments of Prof. Andrade on the slip of metal single crystals, states :—" It is a queer experience to handle such a crystal in the form of a long rod. It is as weak as a similar rod of plasticine and can be stretched at first by the fingers with hardly any resistance. Directly it is distorted by stretching, however, it becomes stronger, and finally, when it is thoroughly maltreated, it assumes the strong rigid character we associate with metal in the ordinary state."

In an aggregate of equi-axial crystal-grains, which is the structure of metals obtained in the ordinary way, by casting or by casting followed by working and annealing, the direction in which slip can occur changes from grain to grain. A fresh start must be made, if the slipping process, begun in one grain, can continue in the next. The necessity of changing direction at each intercrystalline boundary, constituting greater resistance to the applied stress, calls for greater effort in order to effect continuation of movement by slip throughout the polycrystalline mass.

Broadly speaking, metals and alloys may be divided into two classes, *viz.* :—

(1) *Ductile materials, e.g.* copper, silver, pure iron, mild steel, etc., which can be forced to undergo change of shape and dimensions, *e.g.* by rolling or drawing, without fracturing.

(2) *Non-ductile or brittle materials, e.g.* cast iron, hardened high-carbon steel, cast antimony, etc., which have little or no plasticity and undergo an infinitesimally small amount of deformation before fracturing.

In the tensile testing of metals, certain terms are used, which may be defined as follows :—*

Load is the force or combination of forces applied to any part of the machine and is designated by the method of its application.

(a) A *dead load* is steady or invariable, such as the fixed part of a machine or a gradually-applied load as in a testing-machine.

(b) A *live load* is one which is alternately applied and removed, as the weight of traffic passing over a bridge.

(c) A *suddenly-applied load* is one placed on the machine at one instant. It may be applied with or without shock. If the former it may be called an impulsive load.

Strain. Every load, however small, acting on a body, produces an alteration of dimensions or form.

A dead load produces a strain of definite and unvarying amount.

A live load produces a continually varying strain.

A suddenly-applied load produces a strain greater in amount than that produced by the same load steadily applied.

Stress. A metal is composed of atoms, all of which lose electrons easily. It may be thought of as an array of positive ions, with a system of electrons held in common between them all. It is possible to replace one kind of electro-positive atom by another, as long as the newcomer is not incompatible in size and other properties with its neighbours. When the atoms are all of one kind the structure is called a pure metal ; when they are mixed it is called an alloy and it is natural to find that the electro-positive or metal atoms can be alloyed together in a wide range of proportions. The negatively charged electrons which can drift through the structure, between the electro-positive atoms like a cloud, confer on the metal its power of conducting electricity.

There are no direct bonds between atom and atom, binding them all together. The positive ions actually repel each other and are only held in a compact mass by the attraction of the negatively-charged common system of electrons.

This structural arrangement of atoms and electrons affords an explanation of the plasticity of a metal, enabling it to be deformed by hammering, rolling, etc., without destroying its continuity.

The continuous cloud of negative electrons hold the positive atoms together, no matter how the mechanical treatment causes them to slither over each other. A metal is more like a crystallized liquid than a solid.

* *" Mechanical Testing." Batson and Hyde.*

An externally-applied load meets with the resistance offered by layers of atoms to slip over each other and this resistance is called the *stress*. When this resistance is created the metal is said to be in a state of stress. Thus the external load produces strain which calls stress into play.

The determination of the relationship between strain and stress in materials is one of the principal objects of scientific testing.

Strain is measured by a mere number. Load and stress are measured in units of force. It is generally assumed that, in most cases, the conditions which apply to loads, hold also for stresses.

Intensity of stress is the stress per unit area, *e.g.* a stress of P tons acting on an area of A in.^{2} is equivalent to an intensity of stress of $\frac{P}{A}$ tons/in.^{2}

The distribution is uniform when every fraction of the area bears a corresponding fraction of the whole load or stress.

Elasticity is that property by virtue of which a material recovers its original shape and dimensions, when the straining force is withdrawn. A material is said to be perfectly elastic to any external load, if the strain disappears when the load is removed. Such a strain is called the elastic strain.

Plasticity is the property by virtue of which a material may be deformed into various shapes. A perfectly plastic material can undergo an indefinite amount of permanent deformation. Such is "plastic strain." Some materials under an increasing load behave elastically at first, but afterwards begin to "flow." A load has been reached of such value as to produce a strain which persists, when the straining force is removed. Such material is imperfectly elastic at that load and the permanent change of form is called permanent strain or *permanent "set."*

Elastic Limits. The "primitive" elastic limit is the maximum stress which the material can sustain, without undergoing a visible permanent set.

There may be, however, a lack of homogeneity in a material which causes it, at certain points where it is abnormally weak, to undergo minute premature sets. These represent the elastic limit as being lower than the true value.

Test-bars possessing initial curvature will also give false values.

The "natural" elastic limit is the maximum stress which the material will bear, when applied continuously or repeated indefinitely (*e.g.* cyclical variations of stress) without experiencing an increasing strain.

The engineer interprets the behaviour of a given structure from the mechanical properties of its component parts.

Among these properties are tensile strength, compressive strength, torsional strength, shearing strength and the elastic properties associated with all these, as well as ductility, brittleness and resistance to fatigue. Of these, the first-named (tensile strength) is the best understood, the most favoured and most informative. There is considerable latitude permissible in the shape and size of specimen suitable for testing this property. Moreover, most structures are liable to be subjected to tensional forces in service. A tensile test is made by gripping each end of a relatively short length of the material, the dimensions of which have been previously ascertained, and

subjecting it to a pulling force in a direction parallel to its axis. This force is applied and measured in various ways.

The specimen may be round, square or rectangular in section.

The data obtainable in a tensile test are as follows, *viz.* :—

Elastic Limit. Expressed in terms of load per unit area of cross-section. Removal of the load at any value below this limit results in the specimen returning to its previous dimensions. So small are the extensions within this range of stress that special instruments, known as extensometers, are used. These magnify the extensions and make it possible to measure

STANDARD TENSILE TEST PIECES

Abstracted from B.S. 18

FLAT TENSILE TEST PIECE A
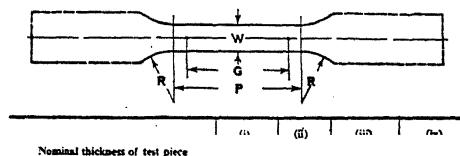
Nominal thickness of test piece

Up to but not including ⅛ in.

		in.	in.	in.	
Width	W	½	1	1½ (max.)	1½ (max.)
Gauge length	G	2	4	8	8
Parallel length (minimum)	P	2½	4½	9	9
Radius at shoulder (minimum)	R	1	1	1	1
Approximate total length		8	12	18	18

ROUND TENSILE TEST PIECES
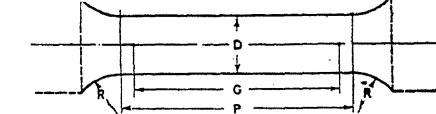
Test	Dia	Cross sectional area	Gauge length	Parallel length (min.)	Radius at shoulder (min.)
	D	A	G	P	R
	in.	sq. in.	in.	in.	in
C	0·564	0·25	2·00	2·25	0·50
Subsidiary standard	0·798	0·50	2·82	3·18	0·70

FIG. 81. STANDARD TENSILE TEST-PIECES.

accurately a movement of the order of 0·0001 in. The longer the gauge-length of the specimen the smaller the extension per unit length (usually 1 in.) which can be accurately measured. When these extensions per unit length are plotted against increments of load, a linear relationship is obtained. At the point where the curve deviates from this straight line, the proportionality of stress to strain (load to extension) as originally discovered by Robert Hooke, comes to an end. This is called the *Limit of Proportionality* and, like the Elastic Limit, is expressed in terms of load per unit of cross-sectional area. It does not appear always to coincide with the elastic limit.

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Hooke's Law. Within the limits of elasticity, stress is proportional to strain.

This relationship refers to all kinds of strain and stress and is exemplified in the stress-strain diagram as a straight line.

$$\text{Stress} = \frac{\text{Load}}{\text{Cross-sectional area}} \quad e.g. \text{ tons/in.}^2$$

$$\text{Strain} = \frac{\text{Extension of gauge-length.}}{\text{Original gauge-length.}}$$

Yield Point is expressed as load per unit cross-sectional area and is the point at which there is a sudden abnormal increase in strain under a uniform rate of loading.

Maximum Load is expressed in tons, pounds or kilograms. This is the load which must be applied, in a normal tensile test, in order to break the specimen.

Ultimate Stress, expressed as a ratio :—

$$\frac{\text{Maximum load}}{\text{Original cross-sectional area.}}$$

This is the tensile strength, as usually required in specifications. In Great Britain it is usually given in tons/in.², on the Continent in kilograms/m.m.², and in the U.S.A. in lbs./in.²

Elongation is expressed as a percentage of the original gauge-length, marked on the test-bar by means of a centre-punch.

$$\text{Percentage elongation} = \frac{L_1 - L}{L} \times 100$$

Where L = original gauge-length and
L_1 = gauge-length after test.

The original length (L) must always be stated in test-reports because different values of L give different percentage elongation results. This is due to the fact that elongation is not uniform along the whole length of the parallel portion of the test-piece. The metal adjacent to the fracture of a ductile specimen elongates more than metal at a distance therefrom. The fractured portion must be included in the result, therefore the greater local elongation at fracture will be bigger in proportion to L, if L be small than if it be large. The following example demonstrates this point :—

TABLE XXA.

Value of L In.	Value of L_1 including fracture In.	L_1—L.	Elongation per cent.
1·0	1·71	0·71	71
3·0	4·32	1·32	44
5·0	6·8	1·8	36

For round sections if the ratio, gauge-length : cross-sectional area, be constant, the per cent. elongation will be constant. Per cent. elongations taken on test-bars in which gauge-lengths and cross-sectional areas are dissimilar will not be comparable, unless the ratio L : $\sqrt{\overline{area}}$ =constant.

A ratio of L =4$\sqrt{\overline{cross\text{-}section\ area}}$ is commonly used in Great Britain.

Adherence to this ratio adds to the cost of preparation of the test-piece, but to ignore it is to render scientific comparisons—of percentage elongations on test-pieces of dissimilar size—impossible.

In Fig. 82 the influence of varying gauge-length, on per cent. elongation, is shown for a steel and also for a lead specimen. A given specimen (S) is seen to have a higher value for elongation than one with longer gauge-length (L).

Other things being equal, percentage elongation and percentage reduction of cross-sectional area are lower, the smaller the length of the parallel portion of a test-piece, whilst yield point and ultimate stress are higher.

This influence of the proximity of the enlarged ends of the test-piece on the part undergoing deformation, can be obviated by making the parallel portion not less than 9 times and the gauge-length at least 8 times, the diameter. Sharp angles and abrupt changes of section in test-pieces should be avoided*, their influence being to lower the ultimate stress, this being particularly true for brittle materials like cast iron and hard steel.
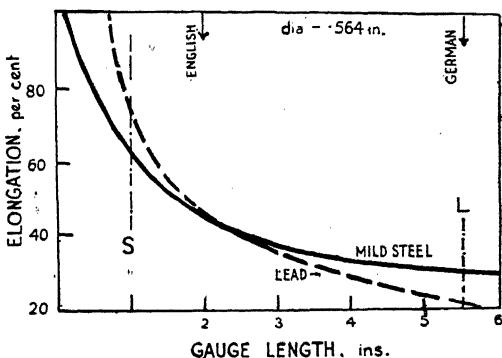
FIG. 82. EFFECT OF GAUGE LENGTH ON PERCENTAGE ELONGATION.

It is essential, in testing brittle materials, to ensure axial loading during the test. This may be done by providing spherical seatings for the shackles (*see* Fig. 83).

Reduction of area, expressed as a percentage of the original area of the cross-section of the parallel portion of the test-bar before testing.

$$\text{Per cent Reduction of Area} = \frac{\text{original area—area at fracture}}{\text{original area}} \times 100$$

Load-Extension Diagram.

A load-extension diagram is shown in Fig. 84 illustrating the behaviour of a mild steel bar when tested in tension. Measurements of the elongation

* *See Fig. 81.*

are taken for equal increments of load, the former being plotted on the horizontal ordinate and the latter on the vertical ordinate. If, at any point on the straight line OP, the load be removed the specimen returns to its original length, O representing zero elongation. Thus the specimen behaves just as an india-rubber band would do. If the load be increased and then released, the specimen no longer resumes its original length (the length under observation being known as the gauge-length), but shows a small *permanent* elongation, *i.e.* the limit of elasticity has been passed, and the point does not lie on the straight line OP when extrapolated as shown. Elongation, now
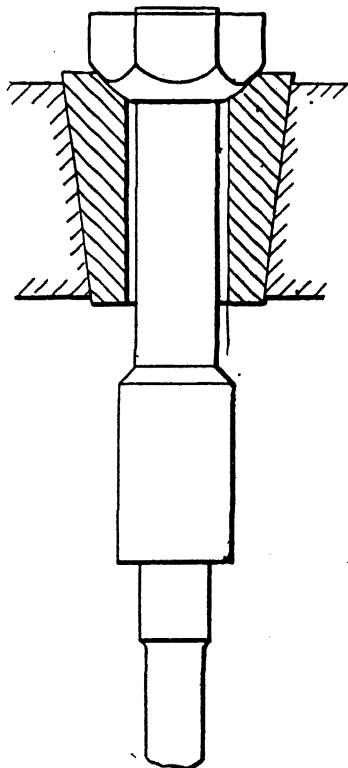
FIG. 83. AN EXAMPLE OF SPHERICAL
SEATING FOR TENSILE TEST-PIECE.

no longer proportional to load, increases at a faster rate than the latter. Point P is the limit of proportionality. A further slight increase in the load to Y results in a marked extension to C, greatly in excess per unit increment of load of any which has previously occurred.

The material has become momentarily plastic, the elongation continuing at constant load, and even under a slightly reduced load in the case of certain low-carbon steels. This is the yield-point and it is often recognizable in lever testing machines, by a sudden drop of the beam, indicating that the

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applied load is in excess of that required to produce the stretching which occurs at this point.

With the exception of certain steels and a very few other metals and alloys, there is no pronounced deflection in the stress-strain curve as described. In its absence, an arbitrary value of stress required to produce a definite permanent elongation is often specified. This is known in America as the yield strength and in Great Britain as the " Proof Stress."
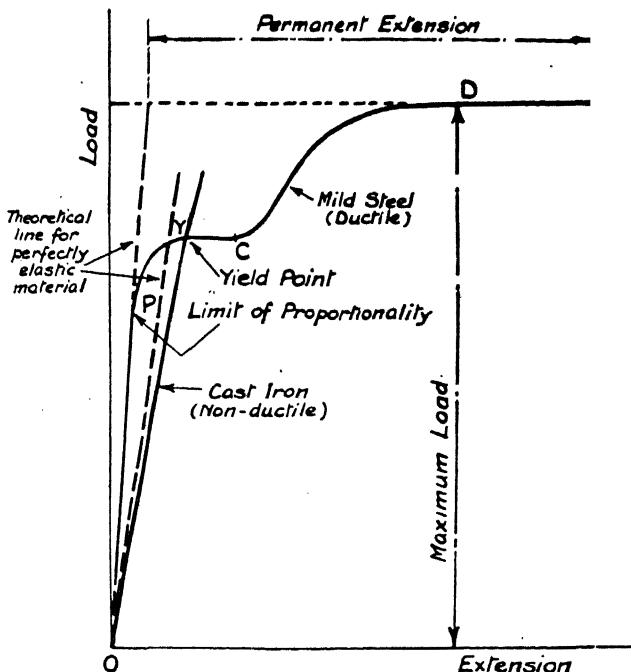
FIG 84. LOAD-EXTENSION CURVES. (*Batson and Hyde.*)

In Fig. 85 OPRS represents the load-elongation curve of a material having no definite yield-point. The limit of proportionality is at P, and OR′ represents an elongation (permanent set.) of 0·10% between gauge-points.

If R′R be drawn parallel to the " elastic " line OP cutting the curve in R, then, with a load OT (corresponding to R) the test-piece has a permanent set of OR′ (0·10%).

Should the proof stress (say ON) give a point on the curve between O and R, the permanent set (OM′) is then less than OR′, *i.e.* less than 0·10%, and the material passes specification. The value of the yield point or proof stress is raised (also the limit of proportionality) by overstraining. That is, if the load be removed after passing the yield point, it will be found, on re-applying the load, that the material has acquired a higher yield-point, which increases to a still higher value after a period of rest at room temperature or after a shorter period at 100° C.

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Testing Machines.

Owing to the inconvenience of handling heavy weights, the simple method of testing by direct loading is now only used in the case of specimens of very low strength.

Loads of much smaller magnitude are increased in effectiveness, by using either a hydraulic press or some form of gearing. Measurement of the load is by calculation from the pressure on the ram or by means of a weighing lever.

Very high loads can thus be applied with the use of very little power. By introducing a lever (single or multiple) between the test-piece and the dead load, a very considerable reduction in the latter, depending on the ratio of the lever arms, can be obtained.
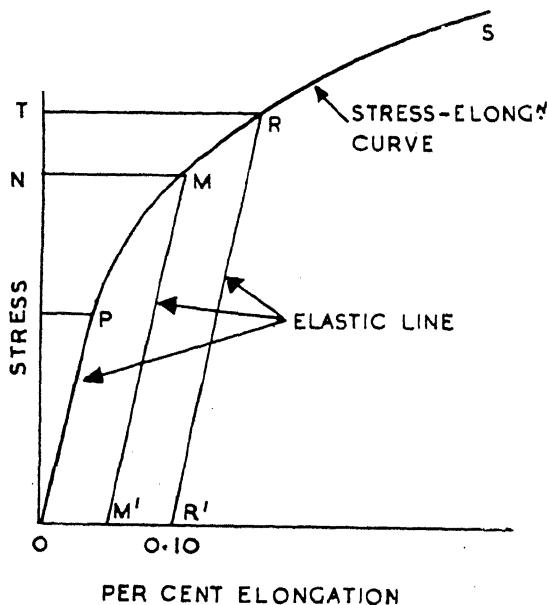
FIG. 85. PROOF STRESS CURVE. (*Batson and Hyde.*)

The following are the chief features required in a testing machine :—

(1) *Accuracy*. This includes ease of calibration and sensitiveness to small differences in the load. In lever machines, wear on the knife edges or slight displacement may result in inaccuracy. On a 3 in. fulcrum distance, a displacement of only 0·03 in. would cause a 1% error in the readings.

(2) *Simplicity*. Parts subject to wear should be reduced to a minimum. Rapidity and ease of manipulation when making commercial tests are very necessary.

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(3) *Accessibility.* Handling and fixing of test-pieces, a clear view of the latter under test, scale and control levers within easy reach, should all be possible to the operator from the front of the machine. Simplicity and convenience for insertion and gripping of test-pieces, attaching autographic apparatus and extensometers.

The vertical machine is more convenient on the whole, but horizontal machines have some advantages, particularly in regard to insertion and gripping of specimens.
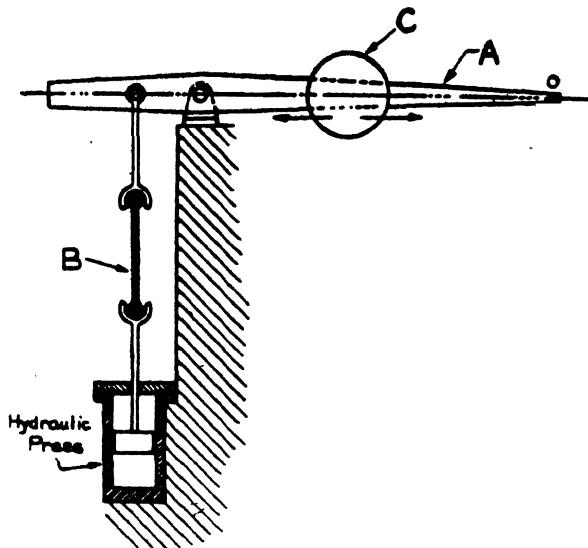
FIG. 86. SINGLE LEVER VERTICAL TESTING
(*Batson and Hyde.*)

The four main parts of modern testing-machines

(1) An arrangement for applying the load.

(2) Means for preventing movement of the end of the test-piece, where attached to the weighing system, at the same time allowing freedom for deformation of the specimen.

(3) Apparatus for measuring the applied load.

(4) Holders (usually called shackles) for connecting one end of the test-piece to part (1) and the other end to part (3).

The load is applied to one end of the test-piece either by (*a*) a straining cylinder, actuated by a pump and accumulator, screw compressor, or oil-pump (hydraulic, rotary or direct driven) ; or (*b*) screw gearing, driven by hand or power (*e.g.* an electric motor). The screw power may be derived from a rotary screw or a rotating nut.

The measurement of the load is effected in British machines commonly by means of a lever A or system of levers with weights or a movable counter-poise C, forming a weighing apparatus (*see* Fig. 86).

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The loading of the lever is effected either (a) by adding separate weights to a carrier placed at a fixed leverage—this is a fixed leverage and variable load type ; (b) by causing a single jockey weight to travel along the beam— this is a fixed load and variable leverage type ; or (c) a combination of (a) and (b).

All three types are illustrated in Fig. 87.

A. Fixed Leverage and Variable Load
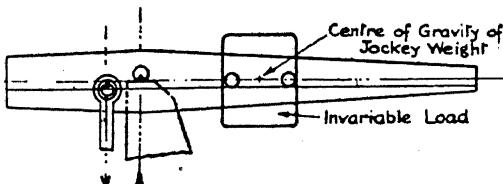
B. Fixed Load and Variable Leverage
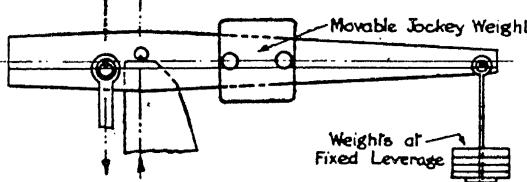
C. Movable Jockey Weight combined with a Load at a Fixed

FIG. 87.   THREE TYPES OF LOADING LEVER.

Fatigue of Metals.

The behaviour of metals under static loading is no real guide to their behaviour when subjected to many repetitions of stress.

The designer must select materials, the properties of which will satisfactorily meet the conditions under which the machine or structural part has to function.   He must consider the effects of stresses caused by loads which frequently fluctuate in value between certain limits, and the name given to the phenomenon of the fragility of metals under these conditions is *"fatigue."*

" Fatigue of metals " is defined as the behaviour of metals when subjected to repeated stresses.

Such stresses are of frequent occurrence in practice and are responsible for the most frequent and serious sources of service failure in cast and wrought metals and alloys.

Alternating stresses will cause a to-and-fro movement along planes of slip, which, although in magnitude less than the total movement required for fracture, for permanent set or even for reaching the primitive elastic limit of the metal, may, nevertheless, if occurring a sufficient number of times, cause failure. Such failure, which may occur after many months, or even years, commences as a minute crack, often originating at a defect, (mechanical or otherwise), spreading further and further until the area of section of sound metal left is insufficient to bear the normal load for which the part was designed. Failure then occurs and examination of the fracture shows two distinct regions (a) that of the crack caused by fatigue and (b) that representing the final sudden rupture. In wrought materials, region (a), if not rendered smooth by the rubbing action of the two opposite faces of the crack, during the period of its formation and growth, will possess a crystalline appearance, wrongly suggesting that the crack has followed an intercrystalline path. The evidence of the coarsely crystalline character of the fatigue-crack in a service-failure has usually been obliterated. The surface is often smooth and discoloured as the result of attrition and penetration of oil and dirt. Region (b) has the same appearance as would be shown if the piece were to be notched and fractured in a normal routine test. This shows that the original grain-structure of the metal has undergone no alteration in service. The crystalline appearance of the fatigue-crack itself is the result of the mechanism of crack development and propagation ; it does *not* indicate that the material has become crystalline in service. To describe the metal as having " crystallized " in service is wrong. As all metals are crystalline anyway, the term has no sort of scientific justification.

Fractures caused by slow, persistent exploitation of slip-planes often present brilliant facets, due to cleavage continuing in one plane across several crystal-grains, instead of changing direction to conform to the changing orientation of the grains, as happens when the metal is fractured in a tensile test.

The coarsely-crystalline fatigue fractures resemble those produced by shock, which acts with such rapidity, that no time is allowed for selection of the most favourably disposed slip-planes in each crystal-grain.

The proportion of service fatigue-crack area to clean fracture area varies from 10 to 95%. The higher this proportion, the higher has been the factor of safety in the design.

An internal imperfection such as a seam, non-metallic inclusion or blowhole often forms the starting point of a fatigue-crack. Surface defects such as scratches, tool-marks, notches, as well as re-entrant angles, abrupt changes in section (" square corners ") etc., act in the same way and are sometimes called " stress-raisers."

Examples of parts subjected to fluctuating stresses are :

(a) steel shafts, supported by brackets and pulleys with belts ;

(b) axles of motor-cars ;

* J. A. G. Stuart (" Mech. World," 1931, 89, 155).

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(c) crank-shafts and connecting-rods of engines ;
(d) teeth of pinions ;
(e) valve springs ;
(f) balance springs of watches ;
(g) pneumatic chisels.

It is clear that the stresses vary in character.

" Repetition " stresses may be sub-divided into :—

(1) Tensile and Compressive Stresses.

(a) *Varying stresses of the same kind* fluctuating between maximum and minimum values (*e.g.* 20 tons/in.^{2} and 10 tons/in.^{2} tension or 6 tons/in.^{2} compression and 1 ton/in.^{2} compression).

(b) *Reversed stresses, i.e.* those which fluctuate from one kind to an opposite kind (*e.g.* 10 tons/in.^{2} tension to 5 tons/in.^{2} compression).

(c) *Alternating stresses, i.e.* those in which the maximum and minimum are equal but of opposite sign (*e.g.* 10 tons/in.^{2} tension and 10 tons/in.^{2} compression.)

(d) The *range of stress* is taken as the algebraical difference between maximum and minimum stresses.

(e) *Endurance* is expressed as the number of cycles of repeated stress required to produce failure.

(f) *Endurance limit* is the highest stress at which an infinite number of reversals may be applied without causing failure.

N.B.—Compressive stress is taken as negative tension, when considering fatigue effects.

(2) Reversed Bending Stresses.
(3) Alternating Torsion.
(4) Repeated Impact.
(5) Reversed Plane Bending.

Wöhler was the pioneer of testing methods for types 1 and 2, the tests being designed to determine the effects of stresses which fluctuate between (a) zero and a maximum ; (b) higher and lower values of the same sign ; (c) tension and compression.

The chief results of Wöhler's experiments were as follows :—

(1) A stress much lower than the ultimate tensile stress and even within the elastic limit may induce fracture if applied and removed or if reduced to a lower value and restored, a sufficient number of times.

(2) The smaller the range of stress variations, the greater the number of repetitions required to produce fracture.

(3) Reduction in the greatest stress increases the number of repetitions required to produce fracture.

(4) When the greatest stress is sufficiently small, it may be applied and removed, reduced and restored or reversed an indefinitely large number of times without producing fracture.

Since an *infinite* number of removals, reductions or reversals cannot be applied, the endurance limits of stress are taken for steel on a 10,000,000 to 20,000,000 basis (removals, reductions or reversals as the case may be).

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Testing Machines.

These may be divided into four main types :—

(a) Rotating cantilever type, like Wöhler's original machine (*see* Fig. 88).

(b) Rotating bar machines with a short length under a constant bending moment (*see* Fig. 90).

(c) Direct stress machines, in which the load is applied by an unbalanced revolving weight.

(d) Direct stress machines, which apply the load by the pull of an electro magnet, excited by an alternating current, *e.g.* the Haigh machine.
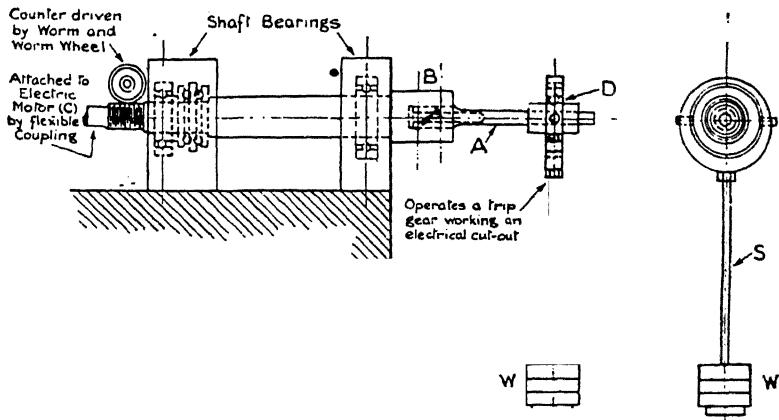
FIG. 88.   ROTATING CANTILEVER FATIGUE-TESTING MACHINE, WÖHLER TYPE.
(*Batson and Hyde.*)

The Wöhler Rotating Cantilever Test.

This applies reversed bending stresses.   A specimen of the section shown in Fig. 88 is placed in the machine and by applying an end load, alternations of stress of a certain magnitude are applied.   The maximum stresses occur at the section AB and are equal and opposite.   Failure occurs after a certain number of reversals of stress have been produced in the rotating specimen.

A second specimen is subjected to a range of stress less than in the first experiment, failure again resulting, but after a longer endurance.   In subsequent tests, the applied range of stress is reduced with each succeeding specimen, until finally a stress is reached under which failure does not occur after, say, 10,000,000 reversals.

Plotting the results (Fig. 89) a graph called the S/N graph is obtained. The fact that the curve becomes more and more nearly parallel to the axis of number of reversals, provides justification for the adoption of 10,000,000 reversals as the basis for the endurance limit.

Table XXI shows the results—obtained on mild steel specimens—from which the graph was drawn.

It may be noted that, with stress limits of $\pm$ 16·3 tons/in.², fracture occurs after only 50,000 reversals whereas the ultimate tensile strength of the steel is about 30 tons/in.²

The endurance limits of stress are $\pm 12 \cdot 75$ tons/in.² — a value much less than half the ultimate tensile strength (static) and probably well within the primitive elastic limit of the steel.

The range of stress corresponding to these limits is clearly 25·5 tons/in.² and this is known as the safe alternating stress range or the " Wöhler safe range of stress." This represents the limiting range of stress, for which the number of repetitions necessary to cause fracture becomes infinite.
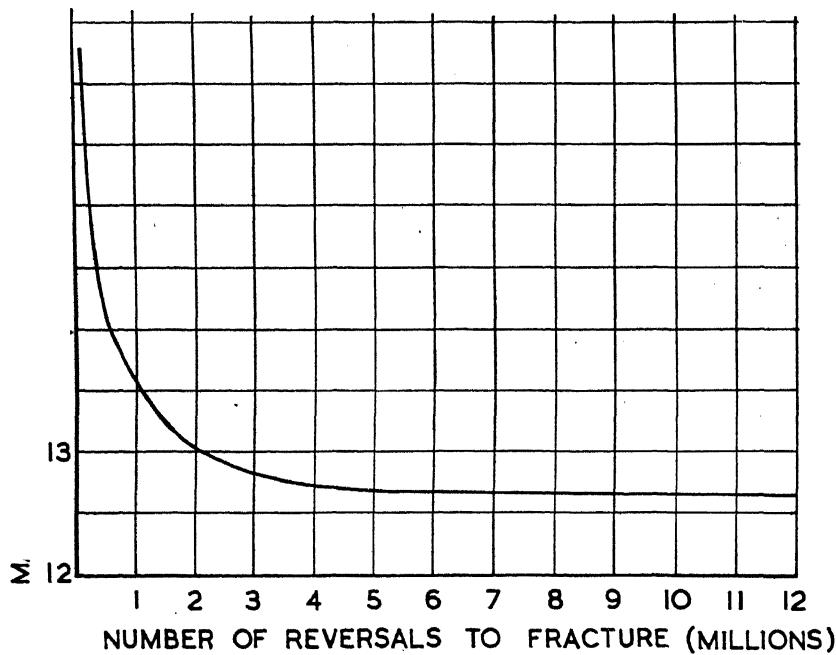
FIG. 89. STRESS-CYCLE (S–N) CURVE.

TABLE XXI.

Limit of Stress Cycle. Tons/in.²	Range of Stress. Tons/in.²	Reversals to Fracture. Millions.	Remarks.
$\pm 16 \cdot 3$	32·6	0·050	Results of tests may be described as
$\pm 15 \cdot 0$	30·0	0·164	follows :—
$\pm 14 \cdot 0$	28·0	0·481	(a) The endurance limits of stress on a
$\pm 13 \cdot 7$	27·4	0·878	10-million reversals basis are $\pm 12 \cdot 75$
$\pm 13 \cdot 4$	26·8	1·394	tons/in.² ; or
$\pm 13 \cdot 4$	26·8	1·264	(b) The limiting range of stress on a
$\pm 13 \cdot 1$	26·2	1·684	10-million reversals basis is 25·5
$\pm 12 \cdot 8$	25·6	2·630	tons/in.² using reversed stresses ; or
$\pm 12 \cdot 8$	25·6	4·537	(c) The limiting range of stress on a
$\pm 12 \cdot 75$	25·4	10·000 specimen unbroken	10-million reversals basis is $\pm 12 \cdot 75$ tons/in.²

The Haigh Test.

Direct tensile and compressive stresses are tested in the Haigh machine, in which stresses are induced by means of electro-magnets excited by alternating currents. Direct alternating stresses, reversed and varying stresses can be tested.

Procedure is the same as in the Wöhler test, a minimum of six specimens being required for any particular range of stress in order to plot an S/N graph.

Fig. 91 shows an S/N graph for a mild steel obtained with the Haigh machine. It is plotted from Table XXII.
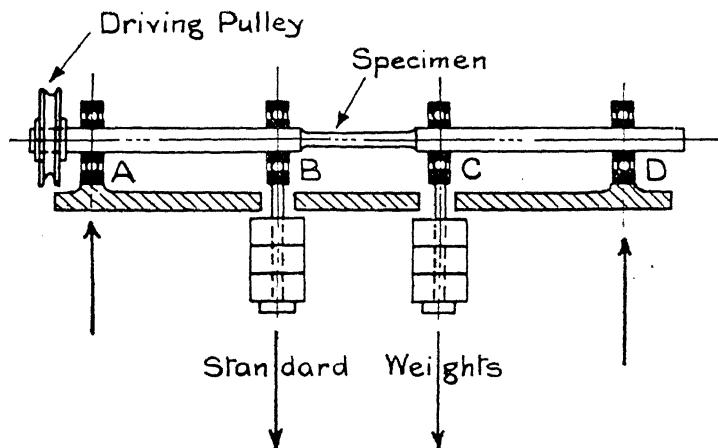
FIG. 90.   ROTATING BAR FATIGUE-TESTING MACHINE, SONDERICKER TYPE.
(Batson and Hyde.)

TABLE XXII.

Ult. Static Stress tons/in.^{2}	Range of Stress tons/in.^{2}	Millions of cycles to Fracture.
26·0	39·3	0·003
—	32·7	0·08
—	31·3	0·056
—	31·3	0·16
—	29·9	0·19
—	29·9	0·80
—	28·8	0·41
—	28·7	1·01
—	27·5	1·37
—	27·5	2·18
—	26·4	1·66
—	26·4	5·90
—	25·4	7·2
		(unbroken)

Table XXIII shows the endurance tests for a number of steels : in each case minimum stresses, which are compressive, zero and tensile stresses are adopted. The first values quoted for each steel are the results for alternating stresses and these are found to agree closely with the results obtained with the Wöhler machine. In all cases the effect of raising the minimum limiting stress is to raise the maximum limiting stress, but to reduce the limiting range of stress.

As regards the equal maximum and minimum limiting stresses, in all cases the ratio of the maximum limiting stress to the tenacity lies between $\frac{1}{4}$ and $\frac{1}{3}$ and the stronger the steel the higher is the limiting range of stress for these alternating stresses, e.g. the limits of stress for the Krupp steel
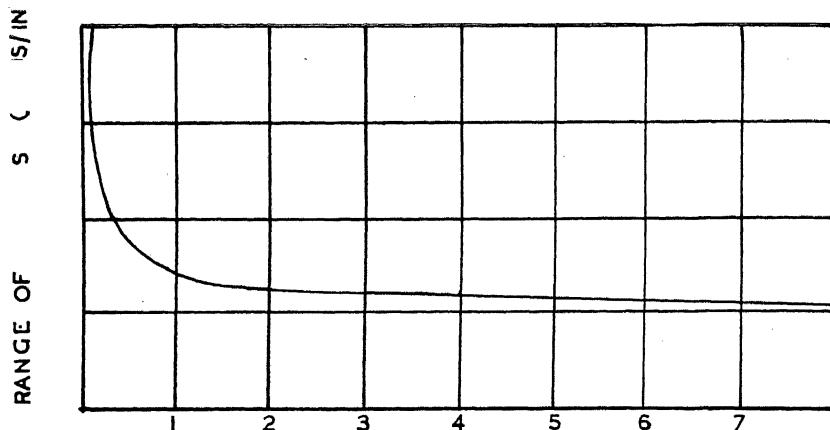
FIG. 91. NUMBER OF REVERSALS TO FRACTURE (MILLIONS) HAIGH TEST. MILD STEEL. EQUAL TENSION AND COMPRESSION.

TABLE XXIII.

Endurance Test Results.

Materials and Ult. Stress.		Minimum Stress (Limiting) Tons/in.²	Maximum Stress (Limiting) Tons/in.²	Limiting Range of Stress Tons/in.²	Ratio of Maximum Limiting Stress to Ult. T. Stress.
Krupp Axle Steel 52 tons/in.²		—14·05	+14·05	28·1	0·27
		0	20·5	20·5	0·39
		17·5	37·75	20·25	0·73
Wrought Iron Plate 22·8 tons/in.²		— 7·15	+ 7·15	14·3	0·31
		0	13·10	13·10	0·57
		11·5	19·2	7·7	0·84
Bessemer Steel 28·6 tons/in.²		— 8·55	+ 8·55	17·10	0·30
		0	15·7	15·7	0·55
		14·3	23·8	9·5	0·83
Steel Rail 39 tons/in.²		— 9·7	+ 9·7	19·4	0·25
		0	18·4	18·4	0·47
		19·5	30·85	11·35	0·79
Mild Steel Plate 26·6 tons/in.²		— 8·65	+ 8·65	17·3	0·33
		0	15·8	15·8	0·59
		13·3	22·55	9·25	0·85

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(T.S. 52 tons /in.²) are $\pm$ 14·05 tons /sq., while the limits of alternating stress for the mild steel (T.S. 26·6 tons /in.²) are $\pm$8·65 tons /sq. in.

Values of maximum limiting stress considerably exceeding the elastic limit are not of practical importance, since stresses which would produce considerable strains cannot be used in machines or structures.

The most important practical relations are those between the repetition limit (minimum stress zero) and the reversal limit (equal and opposite tension and compression).

Over this region the variation of the stress range is not great ; in fact, for a short distance on either side of the alternating stress limits, the range is practically constant.
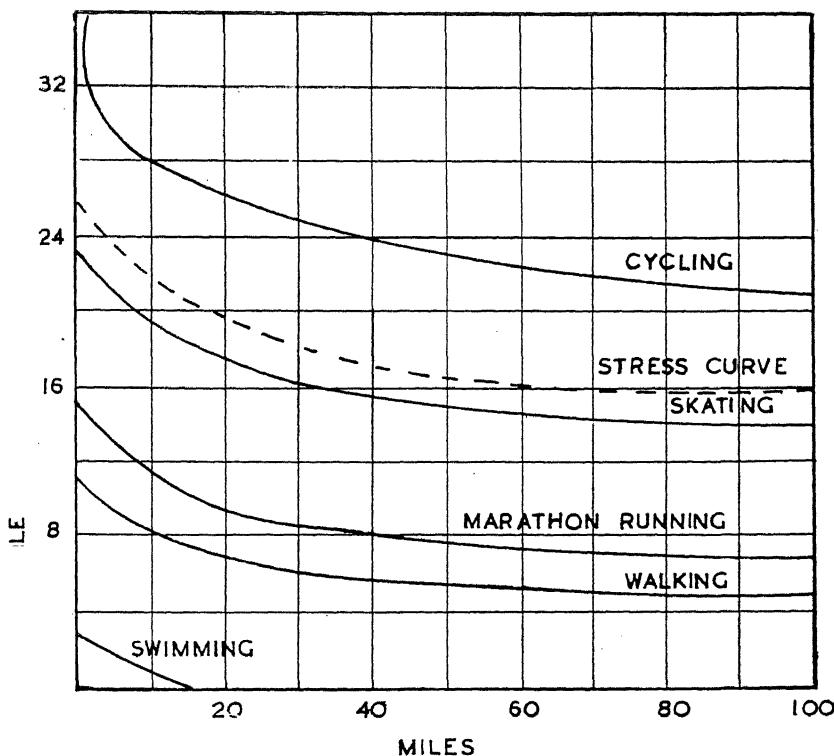
FIG. 92. HUMAN ENDURANCE CURVES.

The " human endurance " curves for various sports in Fig. 92 (Prof. Hele-Shaw) distance v. rate, show a general similarity to the S /N curves.

Thus a cyclist is unable to maintain a speed of 24 m.p.h. for a greater distance than 40 miles. Neglecting the temperament and will-power of the athlete, a limit of " muscular fatigue " is reached, affecting the muscles, and, as a result of a certain effort, these muscles could not be reinvigorated sufficiently quickly to maintain the effort beyond a certain distance.

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Comparison with the S /N curve must not be pushed too far, since factors, other than muscle fatigue, would prevent unlimited endurance in the case of a human " specimen," and these factors play no part in the subject of fatigue of metals.

Speed Effect.

Experiment shows that in all cases the number of repetitions per minute has little or no influence on the limits of stress, *i.e.* the same results are obtained at 350, 1,000 or 5,000 repetitions per minute.

Correlation with other Mechanical Properties. ˙

The only relation which exists is that between the limiting ranges· of stress and tenacity.

Metals of high elastic limit are, in general, the best able to withstand alternations of stress. Fatigue generally shows no great amount of deformation, even in ductile materials.

R. A. MacGregor* enumerates the service conditions which must be taken into account as follows :—

Abnormal or Unexpected Conditions which Predispose Material to Failure by Fatigue.

(1) Material which is weak in its crystalline arrangement, owing to lack of correct treatment or to actually incorrect treatment.
(2) Material, otherwise good, which is too weak for the purpose intended.
(3) Bad design in some small detail, such as the presence of " square corners."
(4) Bad fitting together of two rotating parts.
(5) Bad alignment of rotating or reciprocating parts in their bearings.
(6) Negligence in allowing bearings to score their journals.
(7) Accidental application of one very heavy blow causing small or minute fracture.
(8) Application of unforeseen stresses due to faulty design.
(9) Presence of non-metallic inclusions at the skin of highly-stressed parts.
(10) Excessive and varying temperature gradients.
(11) Unforeseen heavy vibration.
(12) Bad distribution of stress.
(13) High initial stress in material due to some process in manufacture.
(14) Corrosion eating metal away and forming points of excessive concentration of stress.

The case most frequently encountered is that of the " square corner."

(1) *Weak Crystalline Arrangement.*

This is probably the most important of the three metallurgical conditions. Its existence implies one or more of a number of metallurgical errors, *e.g.* :—

High casting temperature.
Slow cooling after casting.
Insufficient forging.
Forging at too high a temperature.
Forging at too low a temperature.
Omission to anneal or imperfect annealing.

* *N.E.C. Inst. of Engrs. and Shipbuilders (Mech. World, 1930, 87, 203 and 263).*

Annealing at too high a temperature.
Heat-treatment at too high a temperature.
Slow cooling after annealing.

The two first mentioned apply to articles as cast. These errors may be corrected by subsequent working, if they are made in the case of cast ingots.

If microscopic examination of the structure of a fatigue failure shows coarsely-crystalline metal, there are good reasons for supposing that the metal had a dangerously low elastic limit.

Once a piece of metal has been stressed beyond its elastic limit, the days of its useful life are numbered, and if kept in service actively stressed beyond its elastic limit it will most assuredly fail. Thus the crystalline arrangement and the elastic limit are very important, when considering fatigue.

(9) Non-Metallic Inclusions.

These are practically unavoidable, but the quality of the steel varies inversely as the quantity of such inclusions, *ceteris paribus*. Their presence, as an intercrystalline network will obviously provide easy paths for the development of a fatigue-crack.

It may happen that such inclusions lie at a highly-stressed point, in which case they will break under stress and provide the minute rupture from which the fatigue-crack starts.

(13) High Initial Stress due to Manufacturing Process.

This case is rare. Certain components may have been bent or flanged to shape without subsequent treatment. Here the metal may be so highly stressed that the application of even a small alternating stress is sufficient to exceed, in one direction, the metal's power of resistance.

(14) Corrosion.

Severe corrosion should be prevented in parts transmitting stresses. Corrosion is more likely to progress in a stressed matrix than in the same material at rest.

Given sufficient time, failure will occur at very low stresses, in the case of commercial carbon steels under conditions favourable to mild corrosion.

E. C. Rollason* cites the following test-results :—

TABLE XXIV.

Influence of Corrosive Media on Fatigue Strength.

Material.	Ult. Tensile Stress Tons/in.^{2}	Endurance limit. $\pm$Tons/in.^{2} at 5×10^7 cycles.		
		Air.	Fresh Water.	Salt Water.
Steel 0·16% carbon	29·3	16·0	8·9	4·0
Steel 3·7% Nickel	40·5	22·0	10·1	7·1
Stainless Iron (12·9% chromium) ..	40·0	24·5	17·0	13·0

' " *Metallurgy for Engineers,*" *p. 17.*

Once fatigue has commenced, there is no cure for it. No amount of heat-treatment can close up or heal a crack, however minute or isolated, even if confined to a single crystal-grain.

There is no test known which will demonstrate the presence of fatigue in the metal adjacent to the crack. There may be simultaneous propagation of other fatigue-cracks in the immediate neighbourhood, but, in spite of that, the unbroken metal close to the cracks will, with one probable exception, still satisfactorily respond to all the original tests and will reveal no new condition produced by the fatigue phenomenon. That exception is the case of a structure which has been severely strained by one heavy blow, prior to commencement of the crack.

RELATION BETWEEN SURFACE CONDITIONS, INTERNAL STRESSES AND FATIGUE

Considerable attention has been given to the problem of fatigue by J. O. Almen.* Fatigue failures result only from tension stresses, never from compression stresses, and any surface, no matter how smoothly finished, is a " stress-raiser."

Fully 90% of all fatigue failures occurring in service or during laboratory and road tests are traceable to design and production defects, and only the remaining 10% are primarily the responsibility of the metallurgist, as defects in material, material specification, or heat-treatment. While this ratio is not a measure of the quality of workmanship contributed by each department, there can be no doubt that the metallurgist has a better appreciation of his responsibility for fatigue failures than have the designer, the engineer, or the man in the production department, for theirs.

The surfaces of repeatedly stressed specimens, no matter how perfectly they are finished, are much more vulnerable to fatigue than the deeper layers. It has long been appreciated that the vulnerability to fatigue increases as the surface roughness is increased, particularly if the roughness consists of sharp notches and more particularly if the notches are oriented at right angles to the principal stress.

The practice of carefully finishing fatigue test specimens and engine parts is, of course, a recognition of this vulnerability in so far as visible marks or scratches are concerned, even down to assuring that the final polishing marks are parallel to the direction of the applied stress. These precautions are known to be effective in increasing the fatigue-strength of the specimens.

However, other influences akin to notches, so far as fatigue vulnerability is concerned, are retained by such specimens, the surface of which, apparently, is highly vulnerable simply because it is a surface, wherein there is an extra hazard not shared by the deeper layers. This extra surface hazard may be due to sub-microscopic notch effects or to the fact that the surface is a discontinuity, since the outer crystal-grains are unsupported on their outer faces. Whatever the reason for surface vulnerability, the evidence of its existence is strong.

The fatigue strength of the most carefully prepared specimens will be increased if a thin layer of the specimen is pre-stressed in compression†

* " The Iron Age." 1943, 151, 65–69, 125–131.
† O. Foeppl. Stahl und Eisen. 1922, 49, 575.

by a peening operation such as peen hammering, swaging, shot blasting or tumbling, or by pressure operation by balls or rollers.

Further evidence of the extra vulnerability of the surface layer is found in the behaviour of specimens having increased strength in a thin surface layer as in thinly carburized or cyanided specimens or in thinly nitrided specimens. Fatigue failures in such specimens also start below the surface and they show greater fatigue strength than the same material when not surface-hardened. A nitrided specimen is probably superior to the other forms of surface-hardening because, in addition to the higher physical properties of the surface layer, this layer is in a state of compression and has, therefore, less notch-sensitivity.

While on the subject of beneficial internal stresses, mention should be made of surface compressive stress obtainable by heat-treatment. By a rapid quench it is possible, through thermal contraction alone, to trap compressive stress in the surface and corresponding tension stress in the core, but this method, although showing some benefit in fatigue, is not so effective as the other methods that have been discussed.

Cold-working increases the hardness of most metals, including steel, at least in the range of low hardness. It usually results in internal stresses of varying degrees and patterns; it alters the physical properties and sometimes fractures the material. With the known sensitivity of materials to fatigue, it is obvious that the problem of controlling cold work must be solved just as has been done in the control of heat-treatment, in order to benefit by the good effects and overcome the evil. Cold-working can be good or bad, depending on how it is done and for what purpose.

Overdose of Nitriding.

Although the usual experience with nitriding is that it greatly improves fatigue strength, it is possible to overdo nitriding just as it is possible to overdo surface stressing by peening and rolling. The high compressive surface stress that results from nitriding must, of course, be balanced by internal tension stress of equal total value. When deep nitriding is applied to light sections the unit internal tension stress may reach dangerous proportions.

It is known that internally nitrided cylinder barrels are more prone to fail by cracking than cylinder barrels that are not nitrided, since the stress due to nitriding is added to the stress from gas pressure. Care must therefore be used in nitriding thin sections, to gauge the depth of the nitrided layer in proportion to thickness of the section being nitrided.

The carburized layer in a carburized part is stressed in compression. Two opposite faces of ½-in. square section specimen were carburized while the other two faces were protected by copper plating. The specimen was quenched and tempered in the usual manner, after which it was split with a saw. The parts are convex on the outer faces, indicating pre-existing compressive stress in these faces. Of interest here is the magnitude of the compressive stress in the carburized layer and the reduced compressive stress, possibly even tension stress, in a thin surface layer. When carburized parts, such as bearing races, wrist pins and gear teeth are ground, it may be expected that the surface will be stressed in tension.

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The residual stress in crankshafts and other parts hardened by induction heating, and probably also in flame-hardened parts, resembles the residual stress in carburized and hardened parts.

With internal stresses of the magnitude existing in the latter, it can be readily understood why carburized parts are prone to warp during heat-treatment, especially if the design is not symmetrical with respect to the internal stresses.

Residual stresses, due to quenching from relatively low temperatures, may reach considerable magnitudes and may be harmful or helpful to fatigue durability, depending on whether the trapped stresses augment or diminish the tension stresses from the applied loads.

Many materials and processes have been graded, and are still being graded, by laboratory tests which are known to have been costly to industry. For example, the fiction that a carburized part should have a hard case to resist wear, and tough core to resist breakage, arose from laboratory impact tests. In these tests, the strength of the part was judged by the number or intensity of hammer blows it would withstand before fracture. Since gear teeth resisted impact fracture in accordance with the physical properties of the core, it seems logical to specify heat-treatments to bring out the best compromise between the imagined requirements of the case and the core. Being compromises, these heat-treatments were not the best for either region.

If, instead of counting the number of impacts or measuring the intensity of hammer blows to produce fracture, the gear tooth had been examined after the first impact, the tooth would have been found bent, and, therefore, ruined, and it would make no difference how many more blows were required to fracture the tooth.

This compromise heat-treatment resulted in reducing the quality of many millions of gears before it was realised that gear teeth fail by fracture and that fatigue failure, for the usual depth of carburization, always originates at the surface of the case. From this evidence it became clear that the heat-treatment should consider the requirements of the carburized case only, and that the properties of the core were relatively unimportant, because in bending and in torsion, the core serves mainly as a stuffing for the case.

Physical Tests.

Several kinds of impact test are still being used and impact specifications appear in many drawings, but no man can explain and substantiate the significance of the test in terms of the service strength of machine parts.

Elongation and reduction of area are carefully measured and are prominent in our specifications, but their meaning, in terms of serviceability of machine parts, is not known. We are told that " brittleness " must be avoided, but no matter how brittleness is defined, it does not explain why this property is necessarily more harmful than ductility. Most machine parts that are plastically deformed are just as surely failed as if they were broken. It is believed that machine parts, generally, must possess relatively high ductility and they must therefore be heat-treated to develop this property. However, when severe dynamic loads are applied, ductility is forgotten and high hardness, that certainly is well within the range of

" brittleness," is specified. Strong fatigue-resistant gear teeth are file hard. Wrist pins, ball bearings, roller bearings, shafts and cams are hard, and they are strong and fatigue-resistant because they are hard.

A gear tooth is just as surely a spring as the coil that actuates a valve. Why, then, must the one be hard and " brittle " and the other be relatively soft and " ductile " ? Why cannot stronger, hardened materials be used ? The answer may lie in the concept of " brittleness." " Brittleness " associated with the hardness obtained by nitriding is not feared. Nitrided surfaces are not notch sensitive because they are stressed in compression.

Notch sensitivity is probably the inability of a non-ductile material to yield locally and thus reduce tension stresses in local highly-stressed regions, such as notches and scratches. The amount of ductility that is required to overcome " brittleness " depends upon the amount of yield that is necessary to reduce local tension stresses. If the surface is sufficiently pre-stressed in compression, local yielding is not required and, therefore, non-ductile materials will not be " brittle." As understanding of " brittleness " is improved, the use of steels at higher hardness will be expanded.

Chapter 6
MECHANICAL PROPERTIES (TESTING)

Methods for testing metals and alloys for toughness and hardness are described in this chapter.

Almen has drawn attention to the fact that the property of toughness is not an indubitable advantage in every case.

There are, nevertheless, many applications in which it is.

The determination of hardness, in the heat-treatment of metals, is essential.

Toughness.

An old conception of toughness was the difficulty experienced in breaking a material by bending, *e.g.* leather. The toughest material was that which would survive the greatest number of reverse bends. This conception leaves out of account the force required to produce the bending, thus assigning, to lead, a higher degree of toughness than to a material which would fail when bent fewer times, but would require more energy to bend it. The criterion of toughness in a material is still often regarded in terms of the difficulty experienced in breaking it by bending. Lead has great plasticity but low strength, whereas a hardened steel tool has little or no plasticity (at room temperature), but great strength. Mild steel possesses intermediate values of both these properties, but the act of bending a mild steel rod once, brings us appreciably nearer to breaking it than does the same operation on a rod of lead of similar size. Less effort is required to effect a given amount of deformation on the more plastic lead which does not stiffen, however many times it may be bent to and fro.

Mild steel, on the other hand, is stiffened by the bending deformation, necessitating an increase in the effort required to increase the angle of the bend or to reverse it. There is a limit to the amount of slip which can occur in a crystal grain of iron, beyond which incipient fracture sets in. This limit in the case of a crystal-grain of lead is reached only after considerably greater deformation.

The conception of toughness is perhaps simpler, when one considers the relative behaviour of specimens of a given material subjected to different kinds of treatment before testing. Compare, for instance, the behaviour of a hardened steel knitting needle with that of an annealed needle. The former would break before bending permanently, whilst the latter could easily be bent through an angle of 90° without breaking. A tough material may possess flow-resistance as well as deformability, both in good measure, but a brittle material possesses only the former, being deficient in the latter. A brittle material may possess high tensile strength, like the hardened steel needle or very little tensile strength like antimony, which can readily be crushed to powder in a mortar.

The Bend Test.

In a bend-test, the specimen is stretched in tension round the outside of the bend and is subjected to compression on the inside of the bend, whilst there is a neutral axis down the centre of the specimen.

When the bending is reversed, the outside, which was formerly in tension, is now subjected to compression and the inside which was formerly in compression now becomes the outside and is stretched in tension. Thus the crystal grains in the plane of bending pass from a maximum tension through zero to a maximum compression in one complete bend, the process being reversed in the next complete bend.

It is thus clear that materials giving a high percentage elongation in the tensile test should bend easily without fracture. Such materials are suitable for press-work operations, e.g. the manufacture of cartridge cases.

In a given operation the material is often required not only to bend at a sharp angle, but to flow readily when passed through the die of the press. Material of low tensile strength would break under the severity of the drawing operation, which necessitates stretching; therefore the ductility of the material must be high and this property implies strength as well as plasticity.

A rough workshop test is made by bending a specimen of the material either by steady pressure or by hammering until an angle of 180° is reached, or until the specimen fractures. In the latter case, the angle at which fracture occurs on the outside of the bend is noted.

Alternating or reverse-bend testing devices are often employed. In one
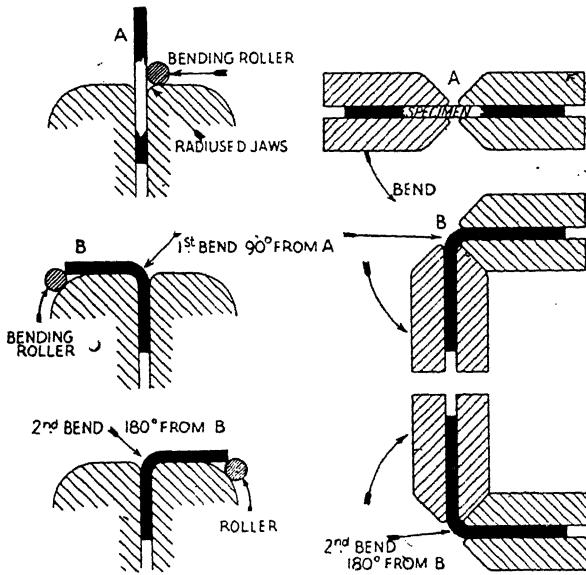
FIG. 93.   DIAGRAM OF TYPES OF BEND-TEST.
(*Hoare and Hedges.*)

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device, one half of the specimen is tightly clamped in a vice with radiused jaws, the projecting portion being bent through a fixed angle (usually 180°), as many times as possible before it breaks.

In another device the projecting portion of the specimen is held in another radiused vice, the movements being made as shown in Fig. 93.*

The Jenkins bend tester, used mainly for testing specimens of mild steel sheet (*e.g.* tinplate) exemplifies the first-mentioned device. This test readily differentiates between specimens cut longitudinally and specimens cut transversely, if the structural condition of the sheet is affected by the presence of " slag streaks " (*see* Figs. 33 and 110).

Other bend-testing machines are available, designed for specific purposes, using hand-power. One such machine which applies the power through
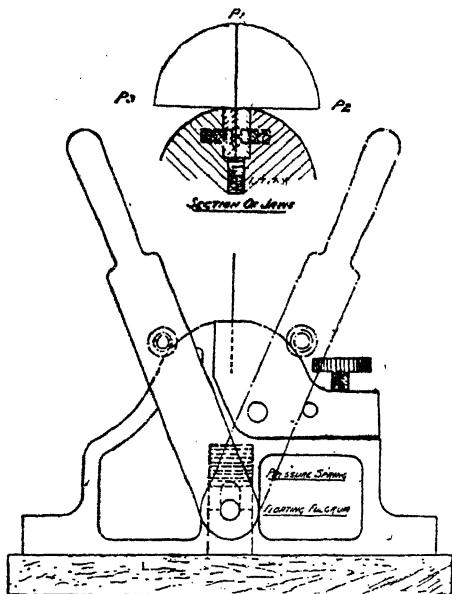
FIG. 94.   JENKINS BEND-TESTING MACHINE.

a train of gears, tests specimens up to 1 in. square. Other devices enable tests to be made in tensile testing-machines. Many machines have been designed in which the test-piece is bent through a small angle under rapidly applied alternating stresses. The Arnold and the Landgraf-Turner alternating impact machines are examples of this type.

In the Sankey hand bending test, the number of bends ($91\frac{1}{2}°$ angle) through which the specimen is bent backwards and forwards and the energy required for each bend are recorded on a chart. " The energy absorbed in breaking the Sankey test-piece is practically proportioned to the product of the average bending moment and the number of bends withstood. These two

* *W. E. Hoare and E. S. Hedges, " Sheet Metal Industries." Nov., 1943, p. 1,896.*

items may be compared roughly to the maximum stress and reduction of area in the tensile test, *i.e.* for conditions of increasing hardness in general, the maximum bending moment increases and the number of bends decreases."

The Upton Lewis toughness testing machine is very similar in principle to the Sankey machine.

Impact Tests.

The single-blow tests, in which notched specimens are used, is now often used in preference to ordinary bend-tests. Material has occasionally satisfactorily passed the ordinary tensile test and subsequently failed in service in a manner which could not be attributed to errors in design. Parts of machinery which have thus failed have, in some cases, been found to have been subjected to shock. Shock-resisting properties and methods of testing them have therefore been subjects for much investigation. Charpy has shown that material which gives good results in the notched-bar bending impact test, shows no superiority over material which gives low values in such tests, when both are subjected to the ordinary static tensile test.*
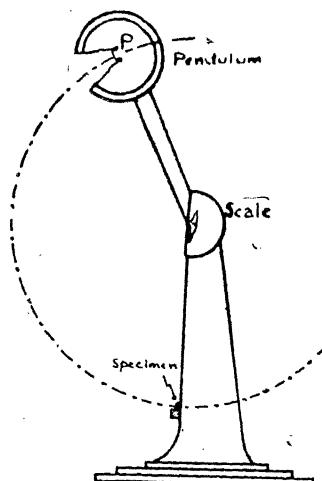
FIG. 95. DIAGRAM OF THE CHARPY
IMPACT TESTING MACHINE.

In the Charpy test, as well as in the Izod test, the test-piece is broken by a single blow from a weighted pendulum, the energy required to fracture the specimen being obtained directly from the loss of energy of the striker.

As shown the Charpy machine (Fig. 95) employs a beam type test-piece and the Izod machine a cantilever type. The same form of V notch is required in the B.E.S.A. specification for both types, although the original Charpy notch was made by drilling a hole 4 mm. diameter in the specimen and sawing through from one side, so as to leave metal to a depth of 15 mm. behind the notch. This is sometimes described as the " keyhole " type of notch. The Izod machine is more used in Great Britain than the Charpy

* *This " notch-sensitivity " is usually due to a difference in heat-treatment.*

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machine. It consists of a heavy base to which are bolted two A-frame standards supporting the pivot of the pendulum (Fig. 96). When released from a standard height the latter, suspended on ball-bearings, on reaching the bottom of its downward swing, strikes the specimen, held in the vice, cantilever fashion. Contact is made with the face of the specimen by a hardened steel knife, fixed to the pendulum. The form of the knife-edge is shown in Fig. 97, in which also are given dimensional details of the specimen and the notch. The energy in the falling pendulum at the moment of striking the specimen is 120 ft./lbs. When the latter is fixed in position, the bottom of the notch facing the striker is level with the top surface of the vice.
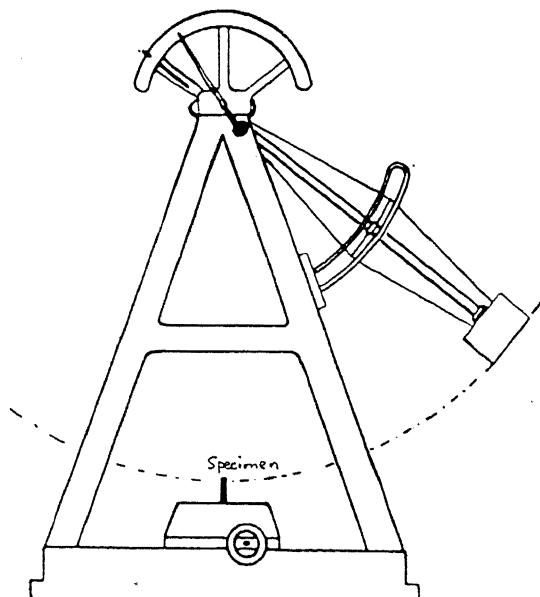
FIG. 96. DIAGRAM OF THE IZOD IMPACT TESTING MACHINE.

Part of the energy in the pendulum is used up in breaking the specimen, so that the former cannot complete the upward swing which it would make if unopposed. A projecting finger attached to the pendulum spindle on the topside of the pivot, engages a pointer and carries it over a quadrant scale, graduated in ft./lbs. Unopposed by any specimen, the pointer is carried completely round the scale to zero. The greater the resistance offered by the specimen to being broken, the shorter the distance travelled by the pendulum in its upward swing, the shorter the distance to which the pointer is carried over the scale and therefore the higher the reading (representing energy absorbed by the specimen) in ft./lbs. Scale-readings are, of course, from left to right.

The distance between the point of contact of striker with specimen and the root of the notch is constant. The concentration of stress which occurs at the notch, compels the specimen to fracture. If it were not for this, ductile specimens would bend only and impede the swing of the pendulum.

Round specimens (0·45 in. diameter) may be used, the size of the notch being so calculated that the results are comparable with those obtained on the standard 10 mm. square section specimens. Round specimens are more economical to machine than square specimens.

Variations in room-temperature when making Izod tests may affect the values obtained.

Sachs & Van Horn* state that the result of a notched-bar impact bending test represents the peculiar quality or tendency of steels to become brittle under the combined action of impact, tri-axial stresses, the stress-concentration induced by sharp notching and low temperatures.

High impact values indicate greater capacity for standing up to the stress-concentration which occurs at a sudden change of section (sharp corners).
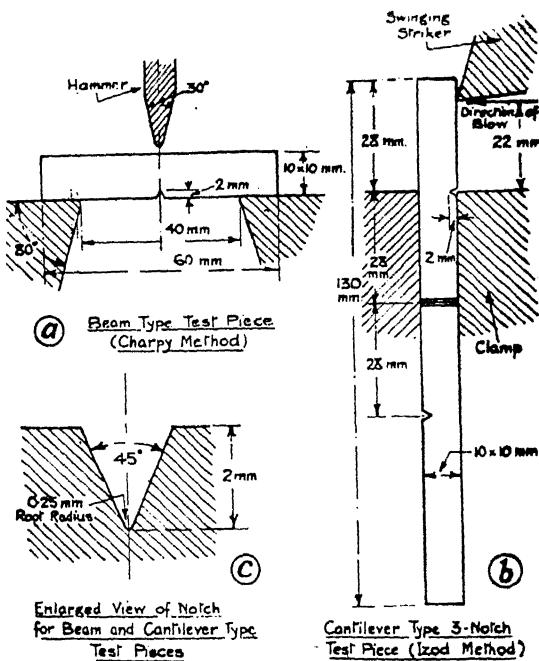
FIG. 97. B.E.S.A. STANDARD BEAM AND CANTILEVER NOTCHED BAR TEST-PIECES.

* " Practical Metallurgy," p. 116.

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Messrs. W. & T. Avery, Ltd., Soho Foundry, Birmingham (makers of the Izod Impact Testing machine) quote the following data for a mild steel of high phosphorus content, viz. :—

Yield Point 22 tons /in.^{2}

Ult. Stress 35·5 tons /in.^{2}

Elongation 30%.

Reduction of Area 47%.

Izod Impact Value 9 ft. /lbs.

This steel would not bend without cracking. A similar steel with low phosphorus content, which would bend without cracking, would give an Izod figure of about 35.

The period elapsing between the inception of a fatigue-crack and final failure by fatigue is related to impact values. Steel, in which the fatigue-crack has spread rapidly, gives low impact values, whereas steel, in which a considerable time has been taken for the crack to spread to an extent which results in rupture, gives high impact values. The information thus derived from impact test-results is of great value in the selection of steels which may be subjected to fatigue. The discovery of a flaw, by routine inspection, will enable the engineer to replace the fatigue-stricken part, before

FIG. 98. PHOTO-MICROGRAPH OF MOTOR-DRIVEN STEEL SHAFT WHICH FAILED BY FATIGUE (150×).
COMPOSITION : CARBON 0·28 %, SILICON 0·15 %, MANGANESE 0·55 %, PHOSPHORUS 0·05 %, SULPHUR 0·04 %, NICKEL 0·50 %

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it is overtaken by disaster. Such latitude would not be provided by material in which a fatigue-crack develops rapidly.

The presence of slag streaks, manganese sulphide and other non-metallic inclusions in rolled steel has a far more adverse effect on impact tests when the direction of the blow, as in the case of test-specimens cut transversely to the direction of rolling, is parallel to the latter, *i.e.* parallel to the long axis of the inclusions.

Sulphide inclusions are shown in the photo-micrograph of a mild steel (Fig. 98). The curves in Fig. 99* show that the direction of the slag-streaks and other inclusions relative to the axis of the test-specimens is not revealed by the ultimate tensile stress, and, although reduction of area% and elongation% figures are affected, the most sensitive test is the impact test, the results of which are lowered even when the test-pieces are cut at an angle of only 20° to the rolling direction.

Figures in Table XXV were obtained by H. R. Philpot.* They show the effect (" temper-brittleness ") resulting from slow cooling nickel-chromium steel from the tempering temperature, instead of oil-quenching therefrom.

TABLE XXV.

Material.	Heat Treatment.	Limit of Proportionality Tons/in.^{2}	Yield Pt. Tons/in.^{2}	Ult. Stress. Tons/in.^{2}	Elongation. %	Reduction of Area. %	Average Izod Impact Value. Ft./lbs.
Nickel-Chromium Steel—							
Bar S.S.1	Tempered and oil-quenched	40·4	47·8	55·5	28·5	64·0	78
Bar S.S.2	Tempered and slowly-cooled	39·6	45·9	54·3	26·5	63·7	9
Nickel-Chromium Steel—							
Bar S.H.1 ..	Tempered and oil-quenched	34·8	44·0	59·3	26·0	61·9	62
Bar S.H.2 ..	Tempered and slowly-cooled	34·8	45·0	60·3	23·0	44·2	8

Hardness Tests.

Although hardness is a physical condition of a metal, it is not a fundamental physical property such as the modulus of elasticity, nor can it be satisfactorily defined. Resistance to indentation by a harder body and resistance to abrasion, although appearing to differ widely in their material requirements, nevertheless have something in common and show a certain amount of agreement in test-results.

Indentation tests are made by causing a hard indenter, such as a hardened steel ball, cone or knife-edge or a diamond pyramid to penetrate the surface of the material under test, subsequently measuring the indentation made.

Abrasion or scratch tests are made by causing particles of the material under test to be torn away from the surface by sliding contact with some other substance, the surface of which remains unimpaired at the end of the test. Such tests should give more reliable indications of wear-resistance and machineability than indentation tests.

Hardness tests present the vexing problem that the actual test itself work-hardens the material under test, thus causing an incalculable increase in the very " property " which is under measurement. But a handy tool or

* *Proc. Inst. Auto. Eng.*, *1917–18*, *12*, *235–370*.
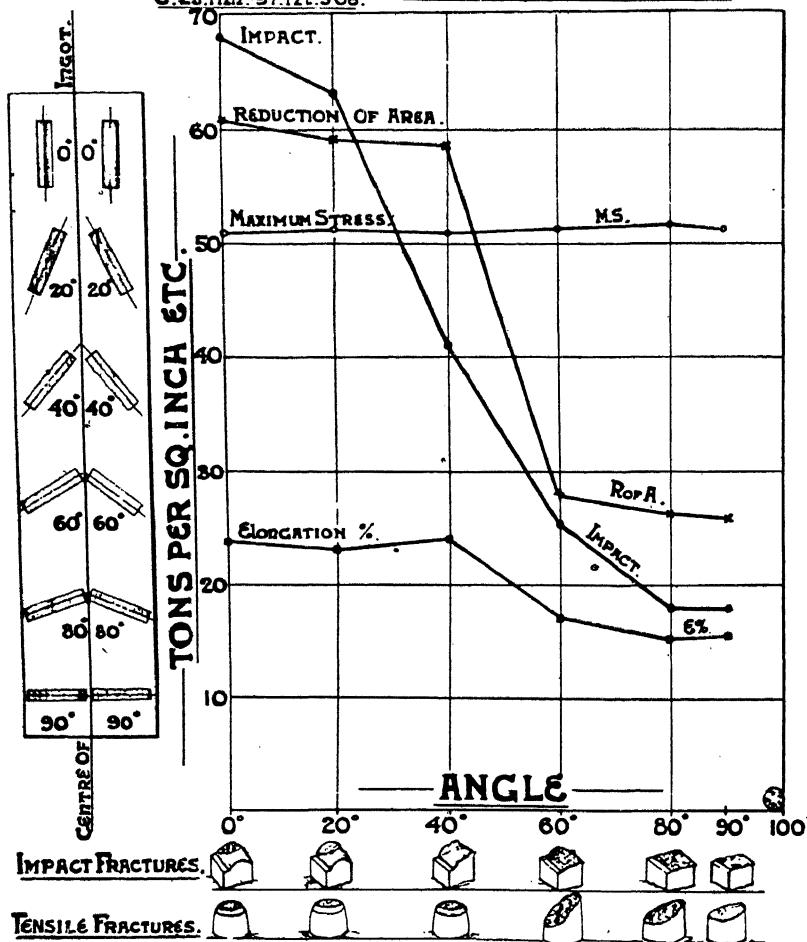
FIG. 99. EFFECT OF ANGLE BETWEEN INGOT AXIS AND SPECIMEN AXIS ON MECHANICAL PROPERTIES. (*Brearley.*)

machine is absolutely necessary, one which gives constant readings when used in testing uniform material. Whatever method be chosen will be governed by such factors as the position occupied by the material in the order of hardness, the size and shape of the part, the condition of the surface, the condition of the material, purpose of making the test, etc.

A form of indentation test which may be applied to all materials, without modification, so as to give absolute values, free from the complications introduced by using an indenter of different material, was suggested by Réaumur in 1722 and modified by Foeppl and by Haigh (1920).

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Foeppl used cylindrical specimens in place of the triangular-section prisms used by Réaumur. Haigh used square-section specimens. When placed at right angles to each other, forming a cross and subjected to pressure, an indentation was made on the surface or along the edge of the specimens.

The measure of this indentation, for a given load, is the measure of resistance to deformation and may be regarded as the nearest approach to a measure of absolute hardness as has yet been devised.

Haigh gave the hardness number as

$\dfrac{L}{R^2}$ where L =the load and
         R =length of damaged edge of specimen.

Such a test as this necessitates the careful preparation of two specimens and cannot be applied to finished articles.

Although various methods were devised for determining hardness employing truncated cones, cylindrical plungers, pyramidal points, knife-edges, etc., it was not until 1900 that a really serviceable instrument was available for the use of engineers and metallurgists. This was the instrument (invented by J. A. Brinell), which is still in use to-day. A hardened steel ball is used as the indenter and is pressed under a known load into the prepared surface of the material to be tested. The diameter of the indentation is measured by means of a microscope with a micrometer eyepiece, and the spherical area calculated. The hardness number is taken as load per unit of spherical area of indentation.

The projected area $\left(\dfrac{\pi d^2}{4}\right)$ would have served equally well.

Brinell No. (H) $= \dfrac{P}{A}$

where P =load in kilograms, A = spherical area of indentation.

D =diameter of ball in mm.
d =diameter of indentation in mm.
h =depth of indentation in mm.
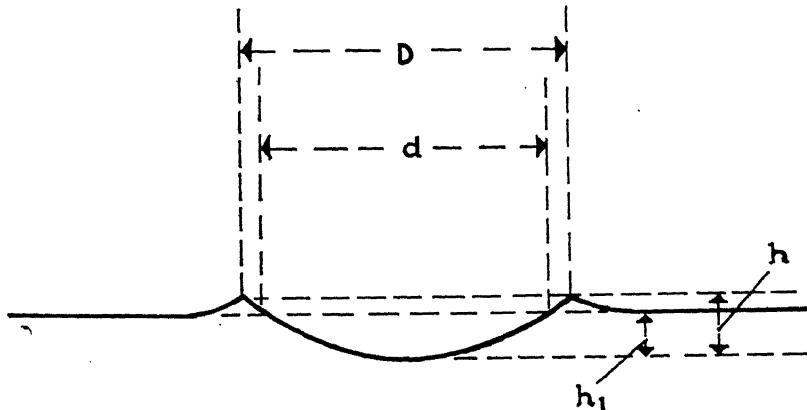
Fig. 100.   Vertical Section of Brinell Indentation.

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Then $h = \frac{1}{2} (D - \sqrt{D^2 - d^2})$

And $A = \pi D h$

$= 1 \cdot 571 D$

A depth-measuring device is supplied with the instrument if desired and, by its aid, the true depth of indentation (h_1) is obtained. Depth h is a false depth, since it is calculated from an indentation, of which D is a false diameter, the measurement of which is made across a circular area formed by a raised lip of metal produced by the penetration of the ball (Fig. 100). The magnitude of this extruding effect varies with different materials. In some cases there is a sinking in of the rim of metal surrounding the indentation.

The hardness numbers based on h are not the same as those based on h_1. The latter are the truer values. Devries has shown that a linear relation exists between load and depth of indentation (h_1) for all loads from 0 to 3,000 kilograms.

The standards as adopted by Brinell are

D = 10 mm.

P = 3000 kilograms for hard metals.

= 500 kilograms for soft metals.

If D be varied, whilst P remains constant, or if P be varied whilst D remains constant, on a given material, the hardness numbers vary.

E. Meyer[*] (1908) showed that these problems are solved if P/D^2 is constant, where P = pressure

d = diameter of indentation

and D = diameter of ball.

Small pieces of metal cannot be satisfactorily tested with a 10 mm. ball and a 3,000 kilogram load. A smaller ball and load are necessary and comparable results are obtainable if the Meyer relationship be maintained.

Thus, if a 5 mm. diameter (D_1) ball be chosen the appropriate load will be

$$\frac{P \times D_1^2}{D^2} = \frac{3000 \times 5^2}{10^2} = 750 \text{ kilograms.}$$

When observing this procedure, very good agreement between the resultant hardness numbers is obtained, even for wide variations in the size of ball, as shown by Baker[†] (see Table XXVI). The relation $\frac{r}{D^2}$ = constant is 30 for steels and hard metals, 10 for copper and 5 for aluminium.

Inaccurate hardness values are obtained if the depth, h, is greater than 1/7 of the thickness of the test-piece[‡] or if the centre of the indentation is less than $2\frac{1}{2}$ times its diameter from the edge of the test-piece.

Penetration of the ball, especially in soft metals, requires time.

From 10 to 15 seconds in the case of hard metals and up to 30 seconds in the case of soft metals should be allowed after the application and before the release of the load.

[*] *Zeit. Vereines Deutsch Ing. 1908, p. 645.*
[†] *Proc. Inst. Mech. Eng. 1918.*
[‡] *In practice, specimens should not be less than $\frac{1}{4}$ in. thick.*

TABLE XXVI.

Comparison of Brinell Numbers, using Balls of different diameters and the Meyer formula.

Steel.	Diam. of Ball. m.m.	Diam. of Indentation. m.m.	Load Kg.	Brinell Hardness No.
A	10	6·3	3000	85
	7	4·4	1470	85
	5	3·13	750	87
	1·19	0·748	42·5	86
B	10	4·75	3000	159
	7	3·33	1470	158
	5	2·35	750	163
	1·19	0·567	42·5	158
C	10	3·48	3000	306
	7	2·43	1470	308
	5	1·75	750	310
	1·19	0·411	42·5	311

The surface of the metal, before testing, should be made reasonably smooth by filing, grinding or machining. To ensure ease and accuracy of measuring the indentation, an even higher degree of smoothness (00 emery) is desirable.

There is a close relationship between Brinell Hardness No. and tensile strength.

Tensile strength (tons/in.^{2}) = C × Brinell H. No. C. varies between 0·206 and 0·230.*

Tensile strengths and hardness values obtained by other testing methods, corresponding to Brinell Hardness numbers, for carbon and low alloy steels are given in Table XXVIII.

The angle formed between the two radii—of the Brinell ball—which meet diametrically opposite points on the circular edge of the indentation, is known as the angle of indentation. Meyer showed that the mean pressure per unit area is constant for a given angle of indentation, whatever the diameter of the ball.

Some of the useful applications of the Brinell test and other tests of a similar nature are as follows :—

(1) For indirect estimation of carbon content of steels, *e.g.* in sorting mixed batches of steel. The steels must be in the same micro-structural condition.

(2) For indirect estimation of tensile strength.

(3) For making tests on finished articles without causing damage thereto.

(4) As a guide to machineability.

(5) As a guide to cutting ability.

(6) For control of heat-treatment and uniformity of its effects.

(7) For ascertaining the hardening effects of cold-working or the softening effects of annealing.

*Dillner, Engineering. 1906, Nov. 9, p. 638.

Brinell Hardness testing machines are made in which the load is applied by oil-pressure and measured by a pressure-gauge or by dead weight control. By using a suitable holder for the ball, the test may be made in any testing machine in which a pressure of 3,000 kilograms can be accurately applied. Unless specially hard alloy-steel balls are used, the ordinary hardened carbon steel ball begins to flatten when testing materials above about 400 Brinell H. No.

In Fig. 101 is shown the standard oil pressure machine of the original Swedish design (Aktiebolaget Alpha). This is made by Messrs. J. W. Jackman and Co., Ltd., Caxton House, Westminster.

The ball is held in a chuck (b) which is attached to the downwardly-acting ram of a hydraulic press, using oil instead of water, The pressure is produced by means of a small hand-pump (a) which, when a hand-operated valve on top of the press-cylinder is closed, forces oil into the latter and causes the ram to move downwards, forcing the ball into the surface of the specimen on the anvil.

The anvil (d) has a spherical seating on the small circular table which can be raised and lowered by the handwheel.(c) There is provided also a deadweight control, consisting of a piston, which is accurately fitted, without packing, in a small auxiliary cylinder (g). Being in communication with the main press cylinder, the intensity of pressure is the same on the piston as on the ram. The maximum pressure is regulated by weights (h) which are suspended
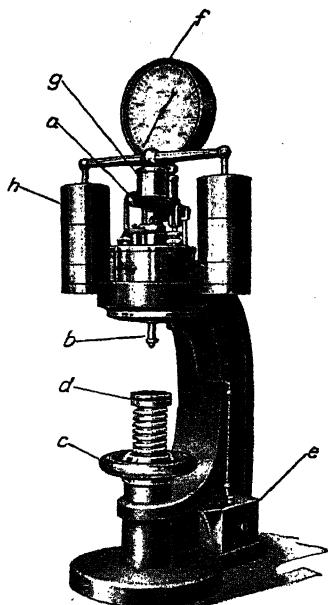
FIG. 101. BRINELL HARDNESS TESTING MACHINE.

from both arms of the cross-bar, carried by the piston. When this pressure is reached, the piston rises in the cylinder and pressure remains constant so long as the piston " floats," forming a small hydraulic accumulator.

The small piston is a hardened steel ball on which the cup-shaped end of the piston-rod presses.

Any leakage of oil past the piston is collected in a receptacle at the bottom of the cast-iron frame. From here it is returned periodically to the reservoir through a funnel fixed near the release valve. By opening the latter, which connects the top of the pressure-cylinder with the reservoir, the pressure is released.

The table is then lowered and the specimen removed for indentation measurement.

As there may be some sticking of the mechanism in the hydraulic, hand-operated machine, the pumping should be done more slowly as the maximum load of 3,000 kilograms is approached, otherwise the latter load may be exceeded.

The Vickers Diamond Pyramid Hardness Test.

The principle of the Brinell test is employed in the Vickers test, viz. :— that of pressing an indenter of hard material into the specimen under test, measuring the resulting indentation and expressing the hardness number in terms of load per unit of area of indentation. The Vickers test, however, differs from the Brinell test in some important details, which are designed to overcome certain inherent disadvantages in the Brinell test. These alleged disadvantages are as follows :—

(1) The inertia and momentum caused by the lifting of the load and its too rapid application.

(2) Variable duration of application of load.

(3) Deformation of the steel ball (most apparent when testing very hard materials).*

(4) Difference in deformation of the materials tested. On a given homogeneous material, different loads give different amounts of deformation and different-sized indentations are geometrically dissimilar. The hardness numbers calculated therefrom are not comparable.

(5) Errors in measurement of indentation.

In the Vickers machine, disadvantages (1) and (2) are eliminated by applying the load very slowly and at a diminishing but always similar rate. Application and removal, after a constant interval are entirely automatically controlled. Inertia and variable duration of load are thus eliminated.

Disadvantage (3) is overcome by using a diamond indenter, which is very much harder than any steel and undergoes negligible deformation.*

* *The hardest metals and alloys can be tested accurately, using diamond (the hardest known substance) as the indenter.*

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Disadvantage (4) disappears, as the diamond is a square-based pyramid in form, therefore the indentations are geometrically similar, whatever their depth, and give comparable results within very wide limits. A cone does this also, but circular indentations lack definition and, at their extreme edge, are difficult and uncertain to read. The square indentation made by a pyramid is easy to measure across the diagonal corners.
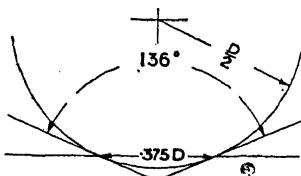
FIG. 102.   DIAGRAM SHOWING DERIVATION OF 136° ANGLE OF VICKERS DIAMOND PYRAMID.

All ball indentations should, if possible be, in diameter, one quarter to one half the diameter of the ball. The mean is 0·375 and the equivalent angularity (*see* Fig. 102) has been taken as standard for the pyramid, viz. :—136°. Hardness numerals, thus obtained, are equivalent to those from ball indentations of ideal size.

$$\text{Hardness No. (V.P.N.)} = \frac{\text{Load}}{\text{Pyramidal Area}}$$

Curve A (*see* Fig. 103) shows that ball numerals obtained with a steel ball loaded to give an indentation of which the diameter is equal to 0·375 times the diameter of the ball are indentical with Vickers Pyramid Numerals (V.P.N.) up to higher hardness values than those obtained under ordinary Brinell standard conditions (curve B).

Disadvantage (5). In place of oblique illumination, the Vickers microscope has vertical illumination which causes the impression to appear uniformly black on a bright background. In place of scale reading which is slow, uncertain and a strain on the eyes, the Vickers method introduces two knife-edges which also appear black. The indentation is " calipered " between these knife edges with great speed and accuracy. The readings are taken from a counter which gives actual figures, outside the ocular micrometer, thus avoiding the possible error of miscounting the divisions on the scale.

The actual V.P.N. is found from a chart or may be calculated by a simple formula.

Additional advantages of the Vickers test are that owing to the smallness of the indentation (about 0·002 in. deep, on hard steels), the " case " and the " core " of case-hardened steels can be tested separately.

The ability to swing the microscope into position, directly centred over the indentation, contributes to the ease and speed with which a test may be made. Specimens with a wide range of sizes can be tested (up to 17 in. deep).

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Round specimens and tubes can be tested, when supported on a Vee-block.

FIG. 103. CURVES SHOWING THE RELATION BETWEEN THE NUMERALS WITH THE VICKERS PYRAMID AND THOSE OBTAINED : (A) WITH A STEEL BALL LOADED TO GIVE AN INDENTATION OF WHICH THE DIAMETER= DIAMETER OF BALL × 0·375, AND (B) WITH A 10 MM. BALL UNDER A LOAD OF 3,000 KGMS.

Application of the Load.

The machine is shown diagrammatically in Fig. 104. It consists of a main frame " F " of V section, carrying the stage " S " and a simple lever " L " of 20 : 1 ratio, applying the load through a thrust rod " Tr " to a tube " T " which is free to reciprocate vertically and carries a diamond indenter " D " at its lower end.

Testing Instructions.

The pedal should first be depressed in order to load the machine. The area to be tested should be smooth and clean, and the work having been placed on the stage, the latter is raised until the surface to be tested just clears the point of the diamond. The small lamp " Lp " may be switched on if necessary to facilitate this adjustment. The starting handle " Sh " is then pressed to release the mechanism, when the test proceeds automatically and terminates with an audible click.

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L

Tr

Sh

T

D

Lp

ₑS

Pl

Tpᵉ

C

Fm

W

F
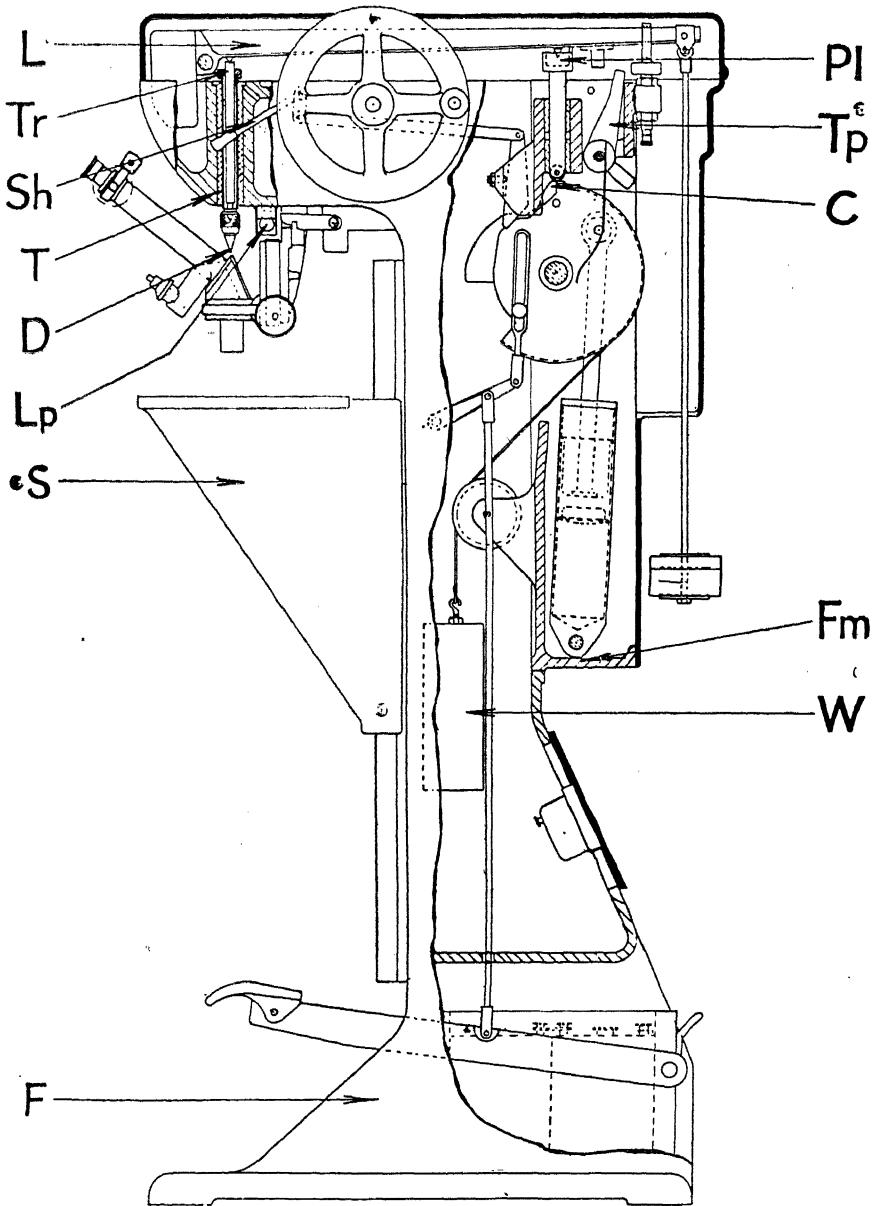
FIG. 104. DIAGRAM SHOWING GENERAL ARRANGEMENT OF VICKERS
HARDNESS TESTING MACHINE.

The stage is then lowered until the microscope can be swung over the work. In focussing the impression, coarse adjustment is obtained by movement of the stage, the fine adjustment being provided on the microscope itself. The measurements are made across the diagonal corners of the square impression.

If, for any reason, the work has not been elevated to clear the diamond within reasonable limits, the test cannot, of course, proceed satisfactorily, and in order to warn the operator of these conditions, a small buzzer comes into operation.

Attached to the main frame is a smaller frame " Fm," which contains all the control mechanism. The plunger " Pl " reciprocates vertically under the influence of a rotating cam " C," its purpose being to apply and release the test load. The cam is mounted on a drum, and when the starting handle " Sh " has been depressed, the whole is rotated by a weight " W " attached by a flexible wire, the speed of rotation being controlled by a piston and dashpot of oil. The rate of displacement of the oil is regulated by an adjustable control valve. The plunger carries a rubber pad at its upper end, which engages with a cone mounted in the beam, thereby ensuring a very slow and diminishing rate of application for the last portion of the load. Since the cam both lowers and raises the plunger, it will be seen that uniformity of loading and duration of the load is attained, all errors due to inertia and premature removal of the load being eliminated. Depression of the foot pedal returns the cam, drum and weight to their original positions. A tripping piece " Tp " supports the beam during this latter operation, and drops out as soon as the plunger returns to its top position. The machine is then ready for another test, and it will be seen that no external power is necessary other than that provided by the operator himself in depressing the foot pedal.

For general purposes, a load of 30 kilos is applied, but as the load may be varied without affecting the numeral, weights are supplied, ranging from 5 to 120 kilos, to suit extremely thin materials and any other special requirements. All impressions are, of course, very small, and the damage to finished work is entirely negligible.

Precautions in using the Vickers Machine.

(1) Apply the test only to smooth, level surfaces, representative of the material.

(2) Select the load with regard to the hardness and thickness of the piece to be tested.

(3) Replace the diamond when it becomes chipped as shown by irregularity of the indentation made.

(4) Check the zero position of the micrometer ocular occasionally.

(5) Adjust the knife-edges of the micrometer accurately perpendicular to the indentation diagonals.

(6) Keep the dashpot full of suitable oil. to prevent impact loading.

(7) Reject all readings—made on a tilted surface—which are indicated by a large difference between the two diagonal readings.

A. F. Dunbar* states that a recent survey of hardness testing in the U.S.A. (the birthplace of the Rockwell testing machine) showed that the

* *Heat-Treating and Forging. Sept., 1943.*

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majority of research institutions used the Vickers machine only, or possessed both types. He further states that :—" The Rockwell test, while ideal for rapid workshop and factory tests, particularly on steel, bears no relationship to the Brinell and Vickers hardness numerals, which, within certain limits, agree very well with each other. Even conversion tables are to be regarded with suspicion and figures for one alloy cannot be used for another. The Rockwell hardness position has got completely out of hand and every year we hear of new scales having to be introduced to cover even harder or softer materials or to fill an existing gap."

Rockwell Hardness Test.

The direction in which this test differs from the Brinell test and its many modifications is one which should recommend it on theoretical grounds, viz. :—the measurement of depth of indentation as distinct from superficial area.

A steel ball (load 100 kilograms) Scale B or a diamond cone (load 160 kilograms), Scale C, is used.

A *minor load* is first applied (10 kilograms). This is done by bringing up the specimen on the elevating table by hand screw until the pointer of the Ames dial is at zero. Under this load, any slack in the machine has been overcome.

The *major load** is now superimposed. This is gradually applied by means of brakes and levers and evenness of application is ensured by brake mechanism inside the main casing. (Fig. 105).

The pointer on the dial moves as the indenter sinks into the test-piece. When it reaches its maximum value, the turning of a crank causes the major load to be withdrawn, leaving the minor load still on. The hardness number is then read directly from the dial. The latter has 100 divisions, each being equivalent to a depth of 0·002 mm.

The surfaces of the specimen must be free from scale and grinding or machining marks. More accurate results are obtainable, the smoother the surface. A flat area should be prepared by filing the surface of round specimens.

Thickness of specimen which may be accurately tested, varies with its hardness. True hardness values are obtained on the hardest steels, if over 0·027 in. in thickness. For softer materials, thicker specimens or lighter loads, or both, may be necessary. Specimens should not show a perceptible impression on their undersurfaces after testing. If they do they are too thin for giving accurate results.

In testing hardened carbon steels, using the cone indenter, the actual depth of penetration is about 0·0027 in. while the effects of the penetration extend to about 10 times this depth, *i.e.* 0·027 in. This should be remembered in testing case-hardened work, especially when the case is thin. The soft subjacent core may cause the indentation to be deeper and too low a reading to be obtained.

* *This is as follows :—A scale 60 kilograms.*
       B scale 100 kilograms.
       C scale 150 kilograms.

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Where only very shallow penetration is possible or permissible, the Rockwell Superficial Hardness Tester may be used. It is used for such materials as nitrided steel, safety razor blades, thinly carburized articles and sheet metal. The regular $\frac{1}{16}$ in. hardened steel indenter is used, but the loads are lighter.

File-Hardness Test.

This is a simple, time-honoured test for " scratch " hardness, capable in the hands of a skilled operator of giving valuable information.

The simplicity and rapidity of the test strongly recommend its application in the control of hardness of surface in production.

In a few seconds, the tester can determine the surface-hardness of the teeth of a gear, various surfaces of a ball or a bearing, etc., without injury thereto.
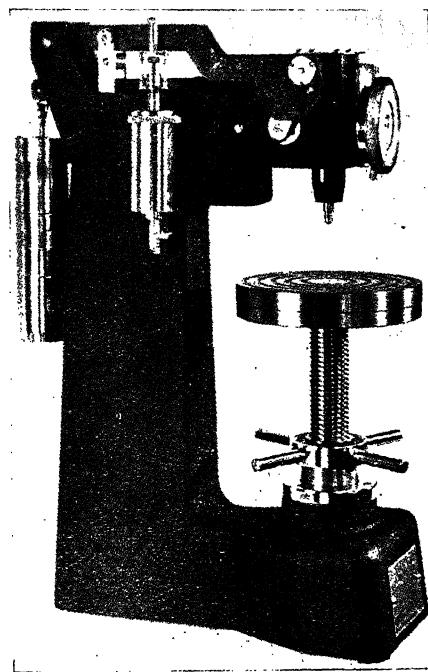
FIG. 105. ROCKWELL HARDNESS TEST.

File-hardness is probably a better criterion of resistance to wear by abrasion—such as gears are required to possess—than of cutting ability in cutting tools. The latter may give excellent service even though less than " file-hard."

Factors influencing the test are :—

(1) Size, shape and hardness of the file used.
(2) Speed of moving the file across the surface under test, also the pressure applied, and the angle at which the file is held.
(3) Composition and heat-treatment of the steel being tested.

File-hardness cannot easily be converted into other well-known hardness values. For example, files which would not cut severely-quenched carbon steel (Rockwell C hardness 64–68), would cut it after tempering at 177° C., although the Rockwell C hardness is still 64–66.

Another instance is that of bastard files which would not cut quenched carbon steel of C. 60 hardness, would only cut steel of C. 58 slowly, but would easily cut chromium steel, tempered at 190° C. of C. 62–64.

Testing-files may only be used a few times. They are most sensitive when they are new, but soon become dull.

The accompanying table shows the rough relationship which exists between file-hardness and Brinell hardness.

TABLE XXVII.*

This Table gives a quick means for determining approximate Brinell Hardness by filing the surface of the metal with a Machinist's New Hand File.

Brinell Hardness.	Ease with which Steel Surface can be filed.
100	File bites into surface very easily. Metal is very soft.
200	File readily removes metal with slightly more pressure. Metal is still quite soft.
300	At 300 Brinell, the metal exhibits its first real resistance to the file.
400	File removes metal with difficulty. Metal is quite hard.
500	File just barely removes metal. Metal is only slightly softer than file.
600	File slides over surface without removing metal. File teeth are dulled.

Bouncing Ball Test.†

A method which can be used in any workshop for testing hardness, depends upon the fact that the harder the surface upon which a hardened steel ball is allowed to fall, the greater the height to which the latter will rebound.

Hardened steel ball bearings are made with a high degree of accuracy and may be used as an inexpensive tool, capable of providing very serviceable data.

A glass tube such as a steam boiler glass, is placed in a vertical position and securely fixed. It is graduated by making file-marks at equi-distant points (Fig. 106).

For calibration, specimens of known hardness are used and the heights of rebound therefrom noted in terms of graduation-marks. The division by equi-distant markings could be confined to the space between the heights of rebound from specimens of lowest and highest known hardness.

* *Heat-Treating and Forging. July, 1942, p. 333.*
† *Heat-Treating and Forging. July, 1942, p. 334.*

Care must be taken to release the ball so that it will be unhindered both in its fall and in its rebound. Although complicated by factors other than true hardness, good results are obtainable on a given class of metal, subjected to different kinds of treatment.
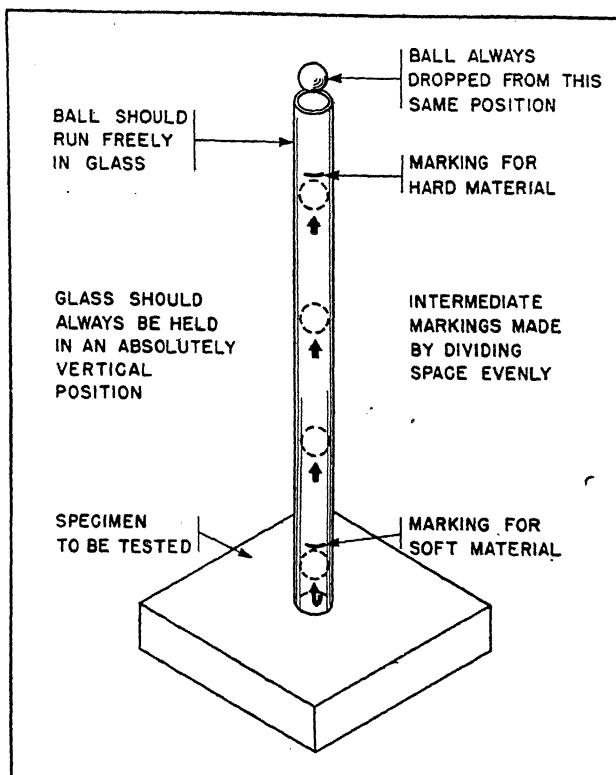
FIG. 106. THIS SKETCH SHOWS THE DETAILS OF THE BOUNCING BALL TEST. THE LOW AND HIGH POINTS CAN BE DETERMINED BY MEASURING THE BOUNCE ON MATERIALS ALREADY MEASURED BY THE FILE TEST OR BY STANDARD MACHINES.

The Scleroscope.

The principle of measuring hardness in terms of the height of rebound of a bouncing steel ball, is employed in the scleroscope. In place of a ball, a small hammer is used, which is allowed to fall by gravitational force, from a fixed height upon the surface of the test-specimen. The hammer falls inside a graduated glass-tube and the height of rebound is read against the scale (Fig. 107).

S. L. Hoyt* describes the scleroscope as a very useful tool especially for check or inspection tests in the shop or laboratory. The primary objective

of using the test for passing or rejecting the material is attained in a reasonably satisfactory manner. The check is not against the true hardness but against a standard (standard block) of essentially similar material processed and tested in essentially the same way. The natural limitations of the rebound principle do not militate against the efficacy of its application under conditions in which the above considerations are recognized.
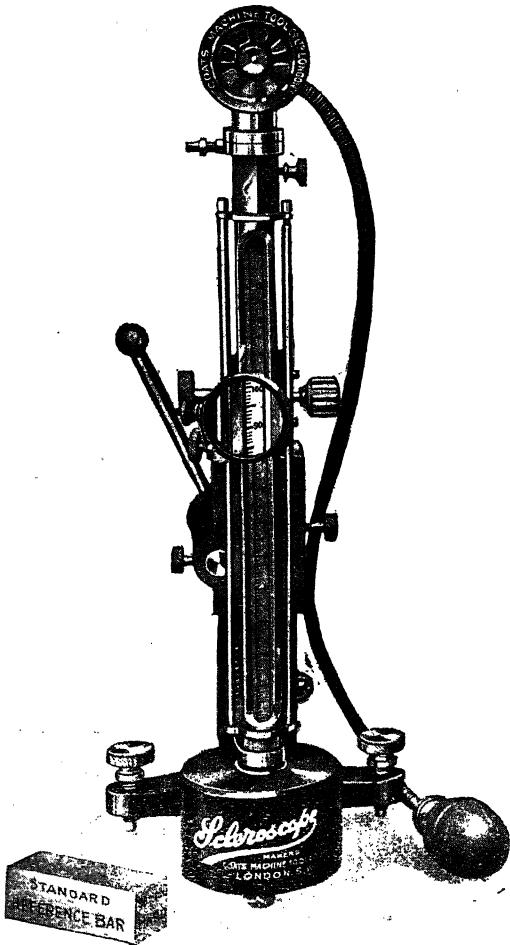
FIG. 107. SCLEROSCOPE. (*Shore.*)

The hammer weighs 1/12 oz., the height of fall is 10 in. The glass tube is graduated into 140 equal parts.

The " Model C " instrument (Fig. 108) is provided with an ingenious device fitted to the top of the glass tube. By this means the hammer is lifted, after making a test, by air-pressure from an india-rubber bulb. It is

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held in position at the top of the tube until released, again by pressing the bulb, for making another test. In striking the surface of the specimen, the point of the hammer makes a small indentation, the work expended in producing which diminishes the height of rebound to a greater extent, the softer the specimen.

When a test is made it is essential that the tube be vertical. Levelling screws and a plumb-rod are provided for this purpose. The bottom of the tube should be held firmly down on the specimen by means of a hand-clamp, when making a test. The surface of the specimen should be smooth, free from scale, grease, etc. If decarburization has occurred the soft skin should be removed. Rough surfaces give low results.

The sound made by the striking of the hammer is a useful guide. A solid thud indicates solid clamping of the specimen and correct rebounds.

A hollow ringing or jingling sound indicates that the specimen is not firmly clamped or does not present a level surface. These conditions give test-results which must be rejected.

There must be no air-space or dirt under thin specimens to cause inaccurate readings.

The diamond-tipped hammer is used for testing steels. The point is slightly spherical and blunt, being about 0·020 in. in diameter. A magnifier hammer is used for soft metals. This has a larger area at the point, thus reducing the depth of penetration, and giving higher rebounds than the diamond hammer. Thus the very low rebounds which the latter would give on soft specimens are " magnified ", are more easily read and enable the significance of small variations in readings to be amplified.

The following precautions in the use of the instrument should be observed, viz. :—

(1) Do not allow the hammer to strike more than once on the same spot, or very near it, as the mechanical hardening caused by the first impact of the hammer will cause a higher rebound from the second.

(2) Mount and clamp specimens as firmly as possible.

(3) Small pieces may preferably be held in the soft steel, copper or lead jaws of a bench vice, using the scleroscope mounted in the swing-arm provided.

(4) The average of a number of readings should be taken, in all cases, unless local characteristics are being studied.

(5) No oil must enter the glass tube. Great care must be exercised in lubricating the operating mechanism. A light high-grade oil should be sparingly used.

The hammer should fit in the tube properly and without friction. Keep the instrument clean at all times.

Causes of Low Readings.

(1) Friction from dirt in the glass tube or on the hammer.

(2) Clogging of the vent-holes in the bottom cap of the barrel, causing interference with escape of air under the falling hammer.

(3) Loose fit of the diamond, as revealed by tapping the stone lightly, when viewed through a magnifying glass.

(4) Instrument out of plumb.

(5) Surface of specimen not level, causing hammer to strike a glancing blow.
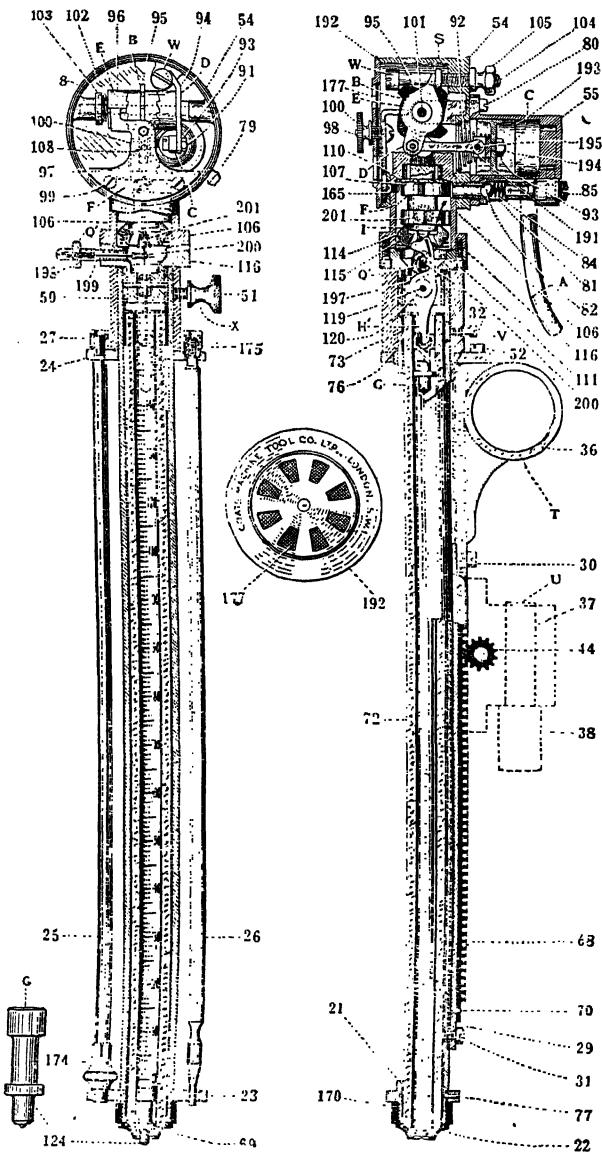
FIG. 108. SCLEROSCOPE, CROSS-SECTIONAL DRAWINGS.

Causes of High Readings.

(1) Under exceptional conditions by wear on the diamond point. Hammers should be redressed and re-standardized.

(2) Entry of air into glass tube chamber, causing the hammer to fall with increased velocity. This is due to the valve ball becoming clogged with dust so that it cannot sit properly.

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Hardness Conversion Table.

The hardness values given in Table XXVIII apply only to steel of uniform chemical composition and uniform heat-treatment and are not applicable to non-ferrous metals nor to case-hardened steels.

TABLE XXVIII.

Approximate Relations between various Hardness Values -and Tensile Strengths of S.A.E. Carbon and Alloy Constructional Steels.

Brinell		Vickers or Firth D.P.N.	Rockwell		Shore Scleroscope No.	Tensile Strength. tons/in.^{2}
3,000 Kg. load. 10 mm. ball.			C 150 Kg. load. 120° Diamond Cone	B 100 Kg. load. $\frac{1}{16}$ in. diam. ball.		
Diam. in. mm.	Hardness No.					
2·05	898	—	—	—	—	196
2·10	857	—	—	—	—	187
2·15	817	—	—	—	—	179
2·20	780	1150	70	—	106	171
2·25	745	1050	68	—	100	164
2·30	712	960	66	—	95	157
2·35	682	885	64	—	91	150
2·40	653	820	62	—	87	144
2·45	627	765	60	—	84	139
2·50	601	717	58	—	81	133
2·55	578	675	57	—	78	128
2·60	555	633	55	120	75	123
2·65	534	598	53	119	72	118
2·70	514	567	52	119	70	114
2·75	495	540	50	117	67	110
2·80	477	515	49	117	65	106
2·85	461	494	47	116	63	102
2·90	444	472	46	115	61	98
2·95	429	454	45	115	59	94
3·00	415	437	44	114	57	91
3·05	401	420	42	113	55	87
3·10	388	404	41	112	54	84
3·15	375	389	40	112	52	81
3·20	363	375	38	110	51	78
3·25	352	363	37	110	49	76
3·30	341	350	36	109	48	74
3·35	331	339	35	109	46	71
3·40	321	327	34	108	45	69
3·45	311	316	33	108	44	67
3·50	302	305	32	107	43	65
3·55	293	296	31	106	42	63
3·60	285	287	30	105	40	61
3·65	277	279	29	104	39	60
3·70	269	270	28	104	38	58
3·75	262	263	26	103	37	57
3·80	255	256	25	102	37	56
3·85	248	248	24	102	36	54
3·90	241	241	23	100	35	53
3·95	235	235	22	99	34	52
4·00	229	229	21	98	33	50
4·05	223	223	20	97	32	49
4·10	217	217	18	96	31	48
4·15	212	212	17	96	31	46

TABLE XXVIII—*cont.*

Brinell		Vickers or Firth D.P.N.	Rockwell		Shore Scleroscope No.	Tensile Strength. tons/in.2
3,000 Kg. load. 10 mm. ball.			C 150 Kg. load. 120° Diamond Cone	B 100 Kg. load. $\frac{1}{16}$ in. diam. ball.		
Diam. in. mm.	Hardness No.					
4·20	207	207	16	95	30	45
4·25	202	202	15	94	30	44
4·30	197	197	13	93	29	43
4·35	192	192	12	92	28	42
4·40	187	187	10	91	28	41
4·45	183	183	9	90	27	40
4·50	179	179	8	89	27	40
4·55	174	174	7	88	26	39
4·60	170	170	6	87	26	38
4·65	166	166	4	86	25	38
4·70	163	163	3	85	25	37
4·75	159	159	2	84	24	36
4·80	156	156	1	83	24	35
4·85	153	153	—	82	23	34
4·90	149	149	—	81	23	33
4·95	146	146	—	80	22	33
5·00	143	143	—	79	22	32
5·05	140	140	—	78	21	32
5·10	137	137	—	77	21	31
5·15	134	134	—	76	21	30
5·20	131	131	—	74	20	29
5·25	128	128	—	73	20	29
5·30	126	126	—	72	—	29
5·35	124	124	—	71	—	28
5·40	121	121	—	70	—	28
5·45	118	118	—	69	—	27
5·50	116	116	—	68	—	27
5·55	114	114	—	67	—	26
5·60	112	112	—	66	—	26
5·65	109	109	—	65	—	25
5·70	107	107	—	64	—	25
5·75	105	105	—	62	—	24
5·80	103	103	—	61	—	24
5·85	101	101	—	60	—	23
5·90	99	99	—	59	—	23
5·95	97	97	—	57	—	22
6·00	95	95	—	56	—	22

Machineability.

Speed of cutting by machine tools is of extremely great importance to the mechanical engineer. Metals and alloys which lend themselves to cutting operations are said to be machineable. A material which is hard to machine from the point of view of the machinist is not necessarily hard as regards resistance to indentation. In actual practice, the Brinell hardness of a heat-treated steel is often a useful, though not infallible guide to its machineability. " Free-cutting " mild steel, which is in demand, where mechanical properties of the product are less important than rapidity of production (*e.g.* on automatic lathes) is more impure than steel whose superior

mechanical properties, especially toughness, are not compatible with maximum " free-cutting " quality. The impurities, *e.g.* sulphur and lead, to which improved ease of machining is due, may actually cause an increase in the Brinell hardness.

Toughness in a metal does not contribute to excellence of machining behaviour, because the cuttings are strong, often spiral-shaped, do not break off readily and are not removed quickly by the stream of lubricant. The short chips obtained from a less tough material are more readily washed out in the case of drilling and boring operations and do not choke the point of the tool.

In most cases, tool-failure is the limiting factor of machineability, which can be measured in terms of the cutting-speed determining a definite tool-life.* That is, machineability may be expressed in terms of tool-life possible without regrinding. With finishing cuts, close adherence to dimensions and the nature of the surface finish left on the work are factors probably as important as tool-life.

Results of lathe-tests made, using high-speed steel tools, the form, size, heat-treatment and composition of which were kept constant, are given in Table XXIX.

The tests were made " dry," feed 0·1115 in. per rev., depth of cut 0·010 in.

The forgings were 8 in. diam. (hollow) with 3 in. diam. holes. The chemical composition was as follows :—

Per cent.

Carbon	Manganese	Phosphorus	Sulphur	Silicon
0·42	0·65	0·02	0·015	0·19

Method of Test.

The test tool was of the broad-nose type and the indicating tool was of the round-nose type.

These two tools were set at equal depths in a special tool-holder at the start of the test. The indicating tool began to cut when the wear on the test-tool was from 0·001 in. to 0·002 in. and this was considered to be the point of failure of the test-tool. In most cases, it coincided with complete breakdown of the tool.

* *T. G. Digges (Mech. World, Oct. 9, 1931, p. ᴊᴛᴘᴛ.*

TABLE XXIX

No. of Forging.	Heat-Treatment			Limit of Proportionality. tons/in.²	Yield Point. tons/in.²	Tensile Strength. tons/in.²	Elongation % in 2 in.	Reduction of Area. %	Rockwell Hardness		Shore Scleroscope Hardness No.	Brinell Hardness No.	Cutting Speed. Ft./min.	No. of tests made.	Average tool life. mins.	Taylor Speed. Ft./min.
	Quenching in water. Temp. °C.	Tempering							B.	C.						
		Temp. °C.	Time hours.													
51	900	260	6·5*	30·8	37·8	55·2	17·8	55·5	100·5	20·6	37	241	180	8	10·5	169
51A	900	427	4*	28	33·8	52·7	20·5	57·8	98·0	15·2	36	229	200	8	7·4	181
51B	900	594	4*	23·4	27	42·75	31·3	64·2	90·0	9·5	25	184	250	7	17·1	246
51C	900	705	4*	21	24·1	39·1	31·5	57·8	87·0	—	27	107	310	8	19·9	310
51D	900	842	4†	21·2	21·7	34·6	34·0	53·5	81·5	—	26	149	325	8	12·0	310

* Cooled in air.   † Slowly cooled in furnace.

Prof. F. C. Lea* states that the type of steel to be machined is one of the fundamental factors in steel-cutting. The best conditions for machining various types of steel are given in Table XXX.

TABLE XXX.

Type of Steel	Best Condition for Machining
Low and medium carbon, cold finished bars.	Lamellar Pearlitic structure.
High Carbon.	Spheroidized (gives longer tool life).
Normal Cold-drawn.	Lamellar Pearlitic structure. Elongation 13–16 %. Reduction of Area 40 %.
Work-hardening alloy.	Hot-rolled and annealed.
Alloy Steels.	Normalized (or annealed).
Carbon case-hardening.	Hot worked.
Shallow hardening fine-grained low alloy.	Lamellar Pearlitic structure (calls for normalizing at higher temperatures).

The composition (straight carbon, alloy, or free-cutting) method of manufacture, *e.g.* as cast, forged, rolled or stamped ; grain-size ; heat-treatment ; mechanical properties ; work-hardening properties ; grain-structure ; and dimensions of piece to be machined are other important factors.

Chapter 7

SUPPLEMENTARY NOTES
ON CARBON STEELS

THE HEAT-TREATMENT OF TOOL STEELS.*

A general outline of the operations involved in heat-treatment of tool steels is given by Schroeder.

The larger the piece of steel to be hardened, the more slowly should it be treated. Where parts are of unequal section, pre-heat to a temperature about 100° C. below the critical point, making sure that thick and thin sections are at the same temperature. The final stage of heating to the hardening temperature may be more rapid. The higher the alloy content, the lower its thermal conductivity, with necessarily longer time to heat to a given temperature.

QUENCHING TEMPERATURE
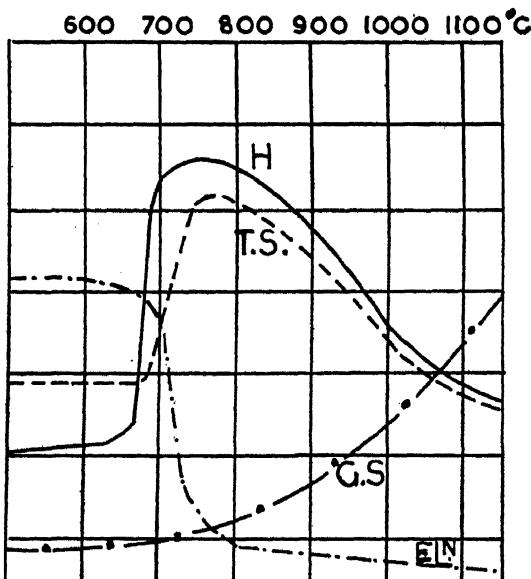
FIG. 109. INFLUENCE OF TEMPERATURE OF QUENCHING ON PROPERTIES
OF 1·0% CARBON STEEL.
H = HARDNESS.   T.S. = TENSILE STRENGTH.   G.S. = GRAIN-SIZE.
ELn = % ELONGATION.

* A. J. Schroeder, " Metallurgia." Dec., 1941, pp. 45–49.

The curves in Fig. 1 show the influence (qualitatively) of quenching temperature on the properties of a 1% carbon steel. When a steel has been heated to the correct hardening temperature, the scale breaks off when water-quenched (frequently also in oil-hardening). If the correct temperature has not been reached, the scale adheres firmly to the steel.

The soaking period for the steel, once it has attained the hardening temperature, varies according to size and structure of the piece.

A reducing atmosphere in the furnace chamber is regarded as essential. When only a portion of the piece is to be heated and hardened, care should be taken to avoid a steep temperature gradient between the heated and non-heated portions.

Of the various types of furnace used for hardening, electric furnaces are not economic unless they are in continuous use and the cost of current does not exceed ⅔d. per Kw.-hour.

For molten baths, the following working temperatures are recommended, viz. :—

Molten lead for temperatures up to 900° C.
Molten Barium Chloride
    (M.Pt. 955° C.)
Molten Sodium Chloride   for temperatures up to 1400° C.
    (M.Pt. 802° C.)
Molten Potassium Chloride
    (M.Pt. 750° C.)

Borax should be added to the barium chloride bath, when used for hardening high-speed steel, to avoid decarburization.

For tempering operations, the following are used :—

For temperatures from 100° C. to 250° C., oil-bath, gas or electrically heated furnaces.

From 200° C. to 600° C., gas, oil or electrically heated salt or metal baths.

For any temperature up to 700° C., electrically-heated, air-circulation furnaces.

Quenching.

The following media are in order of decreasing severity of quench, viz. :—

Mercury (rarely used, except for needle points).

Water containing acids or salts in solution.

*Pure water.

Solution of water and glycerine ($C_3H_8O_3$).

*Mineral Oil (special hardening oil)

Rape Oil.

*Air Blast.

*Still Air

      *Agents most frequently used.

Water should be used at constant temperature (20° C. gives the best results). Higher temperatures favour the formation of steam bubbles and

increase the risk of making scrap. Oil-baths are more fluid at high temperatures, an advantage in facilitating abstraction of heat. Hence the cooling effect is not so seriously reduced by increase in oil temperature (*see* Fig. 34, p. 57).

Interrupted Hardening.

The change from austenite to martensite 'takes place *after* immersion in the quenching liquid, *i.e.* between 300° C. and 180° C. depending on composition of the steel. The rate of cooling should exceed the critical cooling velocity of the steel down to this point, and then slow down. This gives the steel a longer time to undergo the volume change (expansion) associated with the structural transformation, austenite→ martensite. Interrupted hardening is most advantageously applied to water-hardening tools of intricate shape, or widely varying sections. The steel is rapidly cooled in water to about 400° C. (not visibly hot) and then slowly in oil. Quenching stresses are almost entirely eliminated in this way.

Graduated Hardening.

This is effected by quenching in a heated bath (300° C.–180° C.). The tool is plunged from the hardening temperature into the bath and held until it has cooled to the predetermined temperature of the bath. It is then removed and cooled in still air to room temperature (*see* notes on austempering, Chapter I).

Tempering.

(1) 100° C.–200° C. After severe quenching of steels having a high critical cooling velocity, tempering in this range neither softens nor toughens. It does, however, relieve internal stresses and obviates (*a*) subsequent dimensional alterations through natural ageing, and (*b*) subsequent bursting or cracking at changes of section.

(2) 200° C.–300° C. In this range, tempering effects a conspicuous decrease in hardness for all water- or oil-hardening steels (not high-speed nor similar special alloy cutting steels). The higher the tempering temperature in this range, the greater the gain in stability and the greater the loss of hardness.

(3) 350° C.–550° C. Tempering in this range is for high strength, loss of hardness being immaterial. Tools which have to work at high or low temperatures, subjected to impact, thrust or pressure or to rapidly alternating temperatures within the limits of this tempering range, are treated in the same manner as constructional steels intended to resist high stresses.

(4) 550° C.–600° C. Tempering in this range is for hardened high-speed steels. This is because it is desirable to transform the austenite in such steels into harder martensite. If the hardening temperature was too low to produce austenite, the high tempering temperature in this range must not be used. It is doubtful whether high-speed steel in this condition can be serviceably used.

(5) *Control of Tempering Temperature.*

For carbon steels and some alloy-steels, the tempering colour serves to fix, with sufficient accuracy, the required tempering temperature, if within

the range 200° C.–330° C., assuming that heating takes place uniformly and relatively quickly. Also, the surface of the tool must be completely free from grease or salt.

TABLE XXXI.

Approximate Annealing and Tempering Colours.

Annealing Colour.	Temp. °C.	Tempering Colour.*	Temp. °C
Dull red	650	Pale yellow	200
Blood red	700	Straw ,,	220
Cherry red	800	Brown ,,	240
Light red	900	Purple	260
Salmon red	1000	Violet	280
Orange	1100	Dark blue	290
Lemon yellow	1200	Cornflower blue	300
White	1300	Light blue	320
		Blue grey	350
		Grey	400

Steels with high proportions of chromium, tungsten, or cobalt, display no tempering colours such as those in the table, at the temperatures given, if indeed they display any at all. For such steels, temperatures must be determined pyrometrically.

TABLE XXXII.

MAIN REASONS FOR HARDENING FAILURES.

(A) FAULTS IN THE INITIAL CONDITION OF THE STEEL.

No.	Type.	Group.	Fault.	Result.	Remedy.
1	Internal Defect of Material	(a)	Blisters, Shrinkage, Sponginess, Forging or Rolling Faults.	Cracks caused by cooling from the hardening temperature mostly forming large crevices in the direction of grain.	Strict instructions to suppliers as to maintenance of standard of quality. Selection of reliable suppliers of material.
		(b)	Intercrystalline network of cementite.	Serrated or network pattern of cracks, caused during cooling from hardening temperature.	As above. If forged by the hardener, adequate annealing or re-forging.
		(c)	Material "burnt" by annealing.	Soft spots. Shallow hardening. Abnormal strain cracks.	As under (a). If annealed by the hardener, re-annealing.

* It will be observed that on comparison with colours previously given (p. 70) there is some lack of agreement, doubtless due to differences in experimental conditions.

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TABLE XXXII—*cont.*

2	Machining Strain	—	—	Severe distortion. Abnormal strain cracks.	Anneal for stress-relieving.
3	F a u l t y shape of tool	—	Differences in section too great.	Cracks at changes of section. Bursting of edges and corners. Severe distortion.	Radius at corners. Radius at edges. Drill holes through section where thick. When sharp corners are unavoidable, insert iron wire ; this does not result in appreciable softening at the corners. Apply supporting protective strips to prevent sharp edges hardening. Use air-hardening steels.

(B) Faults During Hardening.

No.	Occurrence.	Group.	Fault.	Result.	Remedy.
1	D u r i n g Heating.	(*a*)	I n c o r r e c t h a r d e n i n g temp., too high or too low.	Brittleness or softness. Strain cracks. Poor performance of steel in use.	Control temperature.
		(*b*)	Heating too r a p i d l y . Temperature of core too low.	Cracking of the hardened shell.	Allow temp. to become *uniform*, below hardening temp. Then raise to hardening temp.
		(*c*)	F u r n a c e chamber too small. Piece u n e q u a l l y heated. Wrong adjustment of burner. Heating in blacksmith's forge.	Strain cracks where hot and c o l d e r z o n e s meet. Severe distortion.	Control heat supply. Use a larger heating chamber. Ensure better heat penetration.
		(*d*)	O x i d i z i n g atmosphere in heating chamber.	Decarburization of surface. Soft spots.	Correct burner adjustment.
		(*e*)	Carburization of surface as the result of packing in material too r i c h i n carbon.	Breaking of working edges, in use.	Use packing less rich in carbon.

TABLE XXXII—*cont.*

Nc.	Occurrence.	Group.	Fault.	Result.	Remedy.
2	During Cooling.	(a)	Cooling too rapidly. Cooling too slowly.	Strain cracks. No hardening.	Follow hardening instructions strictly.
		(b)	Adhering steam bubbles.	Soft spots. Strain cracks under certain conditions.	Move tool in quenching bath. Agitate the quenching liquid.
		(c)	Steel enters quench in wrong position.	Severe distortion. Soft spots in engraved dies.	Quench article in direction of long axis. Up and down movement of article in direction of its long axis. Horizontal movement in the case of engraved dies.
3	Scrap resulting from unsuitable design.	—	Cracking or distortion of tools due to severity of quench, when tools are of intricate shape or have parts varying widely in section.		See 2 (c) above. Treat water-hardening steels by means of the interrupted hardening. Treat oil-hardening steels by the graduated hardening (quench in heated bath).

THE TIME FACTOR IN HEAT-TREATMENT OPERATIONS.*

Most of the distortion encountered in the heat-treatment of steel is due to the manner in which the component is heated, rather than to the quality of the steel, improper quenching or machining strains. Steels expand, on heating, uniformly, until the first critical point is reached (the Ac_1 line on the iron-carbon diagram). In the case of hypo-eutectoid steels, contraction begins at this point and continues until the Ac_3 point is reached, beyond which expansion takes its place, and again proceeds uniformly with further rise in temperature.

Thus, if a steel component has a considerable variation in sectional thickness and is heated too rapidly, the thicker sections, when below the critical range, are expanding, whilst medium sections, having reached the critical range, are contracting, and very thin sections, having reached a temperature above the range, are now expanding again. At those temperatures, the yield-point of the steel is low and the internal stresses set up by the simultaneously contending forces of expansion and contraction may exceed it. . A permanent set occurs, and the component is distorted. On cooling, changes in the reverse order take place, the resultant stresses being accentuated by the rapid rate of cooling when quenched. Cracking or even complete rupture occurs either immediately or very shortly afterwards.

* W. Ashcroft, " Metallurgia," Dec., 1941, pp.

In such circumstances, speeding up the operation of heating must be avoided, a slow uniform rate being necessary. The thicker sections should be allowed to soak, especially during the transformation stage, in order to lessen the temperature-difference between thick and thin parts as much as possible.

Although distortion may be minimized by exercising care in heating, it can and does occur as the result of quenching. The difficulty can be greatly reduced by quenching on a falling temperature, especially in the case of high alloy steels. There is always a lag, more marked with alloy steels, which allows of cooling to a temperature which, while still above that of the structural transformation, is a safer temperature for quenching than that suggested by the Ac_3 point.

Most steels have a tendency to grow, when quenched from a high temperature, and to shrink a little when quenched nearer to the recalescence point. By recognizing the importance of this variation and experimenting to find the most suitable quenching temperature, it is possible, with many components, to prevent resultant volume changes almost entirely.

B. Thomas (*ibid.*, p. 44) gives an interesting instance of the effect of lowering the quenching temperature of an S.77 steel (carbon 0·30%, manganese 1·20%)—from the officially recommended 870° C. to 780° C. A marked improvement in impact value (on $\frac{1}{2}$ in. diameter specimens) was effected, and the tempering temperature could be raised from 650° C. to 720° C., making the attainment of the higher impact value a matter of greater certainty in practice.

WARTIME TOOL STEELS.*

Importance of Good Hardening Practice.

The hardening temperature *range* of tool steels is 28°–55° C. (above Ac_3). This does not mean that a given tool steel may be quenched from any temperature within this range with equally good results. It means that small sections should be quenched from the low limit and large sections from the high limit. A good general rule is to use the lower third of the range ·for average sections up to 1 in., the middle third for average sections from 1 in. to 2 in., and the top third for average sections over 2 in.

Soaking.

A conservative rule to cover average soaking conditions after tools or dies—especially irregularly shaped ones—appear to have reached their respective hardening temperatures, would be about 20 mins. per inch of the maximum cross-sectional area of the tool. This does not undersoak the larger sections nor oversoak the small sections.

High speed steel is an exception and should not be soaked, once it has reached the hardening temperature, otherwise undue grain-growth and brittleness ensue.

The ordinary rule of 1 hour per inch of thickness or diameter is alright for tempering or stress-relief heat-treatment.

At hardening temperatures, there is no advantage in soaking after tools have become heated uniformly throughout. Excessive decarburization and scaling, grain-coarsening and brittleness, decreased wear-resistance and even quench-cracking, can result from too long a soaking period.

* H. B. Chambers, " Canadian Metals and Met. Industries," June, 1943.

Quenching.

For all practical purposes, steels commence to change to a martensitic structure, accompanied by expansion, when quenched to the temperature range 205°–260° C. The transformation is completed within the range, 38°–93° C. when the expansion ceases.

If the part be removed from the quenching medium at too high a temperature, it will be softer than if left therein until the transformation is complete. This applies especially to parts of any appreciable mass, *i.e.* thickness of section.

If removed in the danger zone (93°–205° C.) cracking may occur.

Unequal expansions will naturally occur when intricate shapes or relatively large masses are in this zone.

Generally speaking, it is far simpler and safer to quench to just above the temperature of the bath, but not above 93° C. Water and brine baths should be at 21 to 32° C., oil-baths between 38° and 52° C.

An interrupted quench, *i.e.* removing the work from the bath when it has become black followed by air-cooling, minimizes risks of cracking and permits straightening of tools (*e.g.* files) at this temperature if they have become warped. This temperature is below the scaling temperature of steel and very considerably above the temperature at which hardening commences.

Tempering.

This should immediately follow hardening, but never until the dies or tools have approached room temperature. Otherwise the unbalanced quenching stresses in large or intricate sections, or incomplete structural transformations, may cause breakage.

For stress-relief, it is preferable to temper for a longer time at a low temperature than for a shorter time at a higher temperature, provided that the required hardness is retained. It is dangerous to " draw " tools of intricate design rapidly, *e.g.* immersing direct in a salt bath, from the quench. By warming first, stress-relief is set up gradually. One hour per inch thickness is a safe time at temperature for carbon and low alloy steels. 2 hours per inch for high-carbon-high-chromium hot-working and high-speed steels.

Tools which have been insufficiently tempered or stress-relieved are more susceptible to grinding-cracks and also to chipping or breakage in service.

EMERGENCY CARBON STEELS.*

Plain carbon-steels cannot be used under corrosive conditions or at high-temperatures in atmospheres which will lead to loss by scaling or at which creep will occur.

During the emergency created by the war, it has been necessary, however, to substitute carbon-steels for many parts previously made in alloy-steels.

Carbon-steels are much more easily and cheaply produced than alloy-steels, being available in 3 main classes, viz. :—

(*a*) *Finished Steel* comprising plates, sections, bars, sheets, and tinplates ready for fabrication.

* *J. W. Donaldson, " Metallurgia," July, 1942, pp. 94–95.*

(b) *Semi-finished Steel* comprising billets, bars, forgings and stampings, which have to be machined before fabrication.

(c) *Castings*.

In class (b), bars, billets, stampings and light forgings up to 6 in. ruling section are used in all engineering industries, including automobile, aircraft, electrical engineering, marine engineering, and railway and transport work. This class is included in the schedule BS.971, prepared by the Technical Advisory Committee of the Aircraft and Special Steels Committee of the Ministry of Production.   The schedule contains 33 specifications, 9 of which are for carbon-steels, the remainder for alloy-steels.   Based on this schedule, the British Standards Institution has issued a war emergency schedule of wrought steels for general engineering purposes, B.S.970, containing specifications for 58 steels, of which 15 are for carbon-steels and 43 for alloy-steels.

In Table XXXIII, the specifications for ordinary carbon-steels in the T.A.C. schedule are summarized, and in Table XXXIV those for special purpose carbon-steels.

NOTES ON TABLE XXXIII.

T.A.C./1.   Free-Cutting Steel.

Used for rapid machining in automatics and intended for small parts such as bolts, nuts and studs for low duties.   Proof-stress values are higher for this steel in its cold-drawn and reeled condition (up to 20 tons/in.^{2}) but its use is not advised for highly-stressed parts.

T.A.C./2.   Mild Steel.

These steels are for very lightly stressed parts of many descriptions. Material A, although easily machineable, is not intended specifically for rapid cutting, but is suitable for cold work, such as bending and riveting. Material B is suitable for general use, but not for deep-drawing operations. Both A and B can be used for case-hardening.

T.A.C./3.   " 20 " to " 30 " Carbon Steel.

This is for the general engineering construction of such parts as aircraft sockets, tube-plug ends, lightly-stressed brackets, levers, bolts, nuts, studs, etc.   Unless required for service conditions involving bending stresses at sharp corners, its Izod value is unimportant and only maximum stress and elongation need to be specified.   The three conditions in which it can be produced extend its range and multiply the combinations of properties possible.

T.A.C./4.   Bright Carbon Steel.

Is specified to give a tensile strength between 35 and 45 tons/in.^{2}.   It is suitable for automobile and aeronautical engineering parts.   The possible range of composition is wide and includes cold-drawn (bright-drawn) bars which may give a proof-stress up to 30 tons/in.^{2}, as required.   The Izod value is unimportant unless notches or keyways are present in a part designed to take bending-stresses or to be used for aeronautical purposes.

T.A.C.4/N.

Is a free-machining variety of this steel, intended only for nuts.

T.A.C./5. "40" Carbon Steel.

Is a 0·4% carbon-steel suitable for motor connecting-rods and crankshafts, parts requiring strength and wear-resistance without specially high impact-strength; also for bolts, studs, nuts, etc. The wide range of carbon-content specified allows considerable latitude in supplying steels in general use.

T.A.C./6. "55" Carbon Steel.

Whether in the normalized or heat-treated condition, this steel has low impact value. Its applications include cylinders, gears, machine-tool details, hard parts not required to be deeply hardened, such as rifle barrels and certain parts of small arms.breech mechanism. The usual normalizing temperature range is 820° C. to 850° C.

T.A.C./18. Case-Hardening Steel.

Both the steels specified are used, when suitably refined and hardened, for gudgeon pins, tappet rollers, lightly-stressed gears, camshafts, and all relatively lightly-stressed gun and general details requiring high wear-resistance.

NOTES ON TABLE XXXIV.

Of the four special-purpose steels in Table XXXIV, three are vehicle-spring steels for laminated and coiled springs, and the fourth is an engine-valve spring steel. Steels A.O. and A.W. are most generally used for railway vehicles, the choice of the steel depending on whether the spring plates are to be water- or oil-hardened. A.W. is for quenching in water, A.O. for quenching in oil.

Steel A.C. is in general use for coil or spiral springs and is used in the oil-hardened and tempered condition. One of the two main types of steel for aero-engine valve springs is a plain carbon-steel, which acquires its spring-iness from cold-drawing, followed by "blueing" (low-temperature heating to a blue, surface, oxide film). Blooms or billets and wire-rods are ground all over to ensure a flawless surface. A special pickling test is used on specimens cut from the coil, as a check on the possible presence of any surface defects. The "blueing" operation is carried out on the springs after cold-drawing and coiling.

Steel Castings.

Carbon-steel castings usually contain 0·2% to 0·3% carbon and 0·5% to 0·8% manganese, having a tensile strength of 26 to 35 tons/in.² and 20% elongation, as annealed. For certain high-duty castings, the figures are 0·3% to 0·4% carbon, 0·5% to 0·9% manganese, tensile strength 35 to 40 tons/in.² and elongation 17%.

Such steels are used in shipbuilding, some very heavy castings being made for such parts as stems, stern frames, rudder frames, and propeller brackets. In engineering work, turbine casings, crank webs, cylinder covers, piston heads, valve details, etc., are the most important parts made in cast steel.

Steel castings are also used in marine engineering, automobile, aircraft and railway work.

TABLE XXXIII.

Carbon Steels for Bars, Billets, Stampings and Light Forgings.

T.A.C.* Ref. No.	Description.	Condition.	Code.	En. Specification.	Composition %. Carbon.	Silicon.	Manganese.	Sulphur.	Phosphorus.	Proof Stress Tons/in.^{2} 0·1% Proof.	0·2% Proof.	0·5% Proof.	Max. Stress Tons/in.^{2}	Elongation %	Izod* Impact Ft./lbs.
T.A.C./1	Free-cutting Steel	—	—	En.1	0·20 max.	0·10 max.	0·6/1·2	0·2/0·8	0·10 max.	11·5	12	12·5	28 min.	14 min.	—
T.A.C./2	Mild Steel	— —	A B	En.2	0·15 max. 0·15/0·2	0·35 max. 0·35 max.	0·6 max. 0·4/0·8	0·06 max. 0·06 max.	0·05 max. 0·06 max.	11·5 12·5	12 12·5	12·5 13·0	24/28 26/32	27 min. 28 min.	— —
T.A.C./3	"20" to "30" Carbon Steel	Rolled Normalized Oil-quenched and tempered Cold-drawn	A B C D	En.3 En.4 En.5 En.2	0·25 max. 0·15/0·25	0·35 max. 0·35 max.	1·25 max. 0·4/0·8	0·05 max. 0·06	0·05 max. 0·06	11·5 11·6 15 —	12 12 16 —	12·5 12·5 17 —	25/35 25/35 30/40 25/35	25 min. 25 min. 25 min. 17 min.	20 min. 25 min.
T.A.C./4	35/45 Bright Carbon Steel	Bright	—	En.6	0·15/0·4	0·35 max.	0·5/1·75	0·06 max.	0·05 max.	17·5	18	18	35/45	15 min.	20/40 min.
T.A.C./N		Bright	—	En.7	0·12/0·18	0·35 max.	0·5/1·75	0·15 max.	0·06 max.	17·5	18	18	35/45	15 min.	10/20 min.
T.A.C./5	"40" Carbon Steel	Normalized	—	En.8	0·3/0·45	0·35 max.	1·5 max.	0·05 max.	0·05 max.	20	20	20	35/45	20 min.	20 min.
T.A.C./6	"55" Carbon	Normalized Oil-quenched and tempered	A B	En.9 En.10	0·5/0·6	—	0·4/0·75	—	0·05 max.	23 29	23·5 30	24 31	45 min. 55 min.	18 min. 15 min.	— —
T.A.C./18	Carbon C.H. Steel	Refined and Hardened		En.32 En.32	0·18 max. 0·2 max.	0·35 max. 0·35 max.	0·9 max. 0·4/1·0	0·05 max. 0·07 max.	0·05 max. 0·07 max.	15 15	16 16	17 17	32 min. 32 min.	20 min. 20 min.	40 min. 40 min.

* Technical Advisory Committee of the Aircraft and Special Steels Committee of the Ministry of Production.

TABLE XXXIV.

Carbon Steels for Special Purposes.

T.A.C. Ref. No.	Description.	Code.	Condition.	En. Specification.	Composition %. Carbon.	Silicon.	Manganese.	Sulphur.	Phosphorus.	Max. Stress Tons/in.^{2}
T.A.C./25	Vehicle Spring Steels	A.O. A.W. A.C.	Oil-hardened and tempered Water-hardened and tempered Oil-hardened and tempered	En.42 En.43 En.44	0·75/0·90 0·45/0·65 0·90/1·20	0·4 max. 0·3 max. 0·3 max.	0·35/0·70 0·5/1·0 0·45/0·70	0·05 max. 0·05 max. 0·05 max.	0·05 max. 0·05 max. 0·05 max.	— — —
T.A.C./26	Engine-valve Spring Steel	—	Hard-drawn and "blued"	En.49	0·70/0·80	0·35 max.	1·0 max.	0·04 max.	0·04 max.	95–120
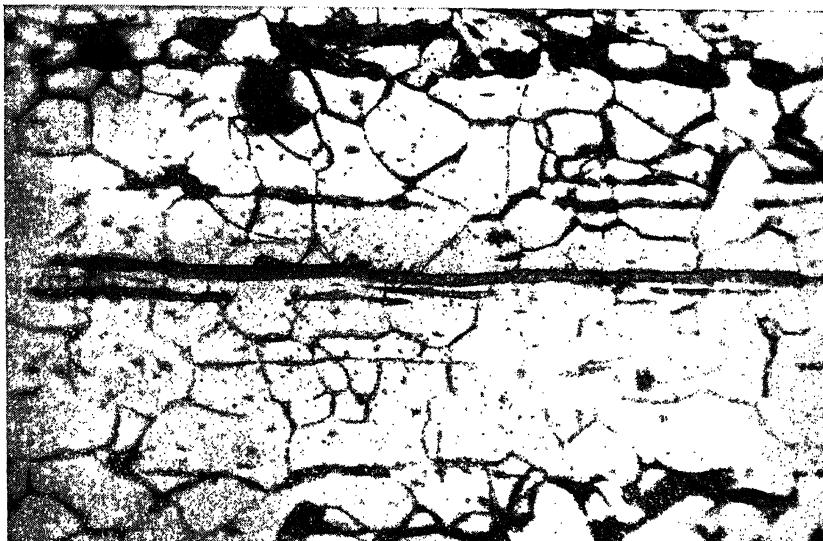
FIG. 110. MILD STEEL (×400), SHOWS NON-METALLIC INCLUSIONS
(GREY, ELONGATED).
TYPICAL OF STRUCTURE OF FREE-CUTTING MILD STEEL.

APPENDIX.

TABLE XXXV.

AMERICAN STANDARD SPECIFICATIONS.

S.A.E.* CARBON STEELS.

S.A.E. No.	Title of Specifica- cation.	Chemical Composition.				Corre- sponding A.I.S.I. No.
		Carbon, per cent.	Mangan- ese, per cent.	Phos- phorus, per cent.	Sulphur, per cent.	
				Max.	Max.	
1008	Carbon Steel	0·10 max.	0·30–0·50	0·040	0·050	C1008
1010	,, ,,	0·08–0·13	0·30–0·50	0·040	0·050	C1010
1015	,, ,,	0·13–0·18	0·30–0·50	0·040	0·050	C1015
X1015 1016	,, ,,	0·13–0·18	0·60–0·90	0·040	0·050	C1016
1020	,, ,,	0·18–0·23	0·30–0·50	0·040	0·050	C1020
X1020 1022	,, ,,	0·18–0·23	0·70–1·00	0·040	0·050	C1022
1024	,, ,,	0·20–0·26	1·35–1·65	0·040	0·050	C1024
1025	,, ,,	0·22–0·28	0·30–0·50	0·040	0·050	C1025
1030	,, ,,	0·28–0·34	0·60–0·90	0·040	0·050	C1030
1035	,, ,,	0·32–0·38	0·60–0·90	0·040	0·050	C1035
1036	,, ,,	0·32–0·39	1·20–1·50	0·040	0·050	C1036
1040	,, ,,	0·37–0·44	0·60–0·90	0·040	0·050	C1040
1045	,, ,,	0·43–0·50	0·60–0·90	0·040	0·050	C1045
1050	,, ,,	0·48–0·55	0·60–0·90	0·040	0·050	C1050
1052	,, ,,	0·47–0·55	1·20–1·50	0·040	0·050	C1052
1055	,, ,,	0·50–0·60	0·60–0·90	0·040	0·050	C1055
1060	,, ,,	0·55–0·65	0·60–0·90	0·040	0·050	C1060
X1065 1066	,, ,,	0·60–0·71	0·80–1·10	0·040	0·050	C1066
1070	,, ,,	0·65–0·75	0·70–1·00	0·040	0·050	C1070
1080	,, ,,	0·75–0·88	0·60–0·90	0·040	0·050	C1080
1085	,, ,,	0·80–0·93	0·70–1·00	0·040	0·050	C1085
1095	,, ,,	0·90–1·05	0·30–0·50	0·040	0·050	C1095

*S.A.E. *Society of Automotive Engineers.*

X The numbers which are prefaced with a cross indicate some departure from regular chemistry for such steels. The number bracketed with the number having a X is the new S.A.E. number. This method of dual numbering will be retained for a suitable transition period to permit steel purchasers to adopt the new numbers with the least inconvenience to specifications, drawing, etc.

TABLE XXXVI.
S.A.E. Free-Cutting Steels.

S.A.E. No.	Title of Specification.	Chemical Composition.				Corresponding A.I.S.I. No.
		Carbon, per cent.	Manganese, per cent.	Phosphorus, per cent.	Sulphur, per cent.	
Bessemer:						
1111	Free Cutting Steel	0·08–0·13	0·60–0·90	0·09–0·13	0·10–0·15	B1111
1112	do.	0·08–0·13	0·60–0·90	0·09–0·13	0·16–0·23	B1112
X1112} 1113	do.	0·08–0·13	0·60–0·90	0·09–0·13	0·24–0·33	B1113
Open Hearth:						
1115	Free Cutting Steel	0·13–0·18	0·70–1·00	Max. 0·045	0·10–0·15	C1115
X1314} 1117	do.	0·14–0·20	1·00–1·30	0·045	0·08–0·13	C1117
X1315} 1118	do.	0·14–0·20	1·30–1·60	0·045	0·08–0·13	C1118
X1330} 1132	do.	0·27–0·34	1·35–1·65	0·045	0·08–0·13	C1132
X1335} 1137	do.	0·32–0·39	1·35–1·65	0·045	0·08–0·13	C1137
X1340} 1141	do.	0·37–0·45	1·35–1·65	0·045	0·08–0·13	C1141
1145	do.	0·42–0·49	0·70–1·00	0·045	0·04–0·07	—

TABLE XXXVII.
S.A.E. Manganese Steels.

S.A.E. No.	Title of Specification.	Chemical Composition.					Corresponding A.I.S.I. No.
		Carbon, per cent.	Manganese, per cent.	Phosphorus, per cent.	Sulphur, per cent.	Silicon, per cent.	
1320	Manganese Steel	0·18–0·23	1·60–1·90	Max. 0·040	Max. 0·040	0·20–0·35	A1320
1330	do.	0·28–0·33	1·60–1·90	0·040	0·040	—	A1330
1335	do.	0·33–0·38	1·60–1·90	0·040	0·040	—	A1335
1340	do.	0·38–0·43	1·60–1·90	0·040	0·040	—	A1340

TABLE XXXVIII.
S.A.E. Nickel Steels.

S.A.E. No.	Title of Specification.	Chemical Composition.						Corresponding A.I.S.I. No.
		Carbon, per cent.	Manganese, per cent.	Phosphorus, per cent.	Sulphur, per cent.	Silicon, per cent.	Nickel, per cent.	
2315} 2317	Nickel Steel	0·15–0·20	0·40–0·60	Max. 0·040	Max. 0·040	0·20–0·35	3·25–3·75	A2317
2330	,, ,,	0·28–0·30	0·60–0·80	0·040	0·040	0·20–0·35	3·25–3·75	A2330
2340	,, ,,	0·38–0·43	0·70–0·90	0·040	0·040	0·20–0·35	3·25–3·75	A2340
2345	,, ,,	0·43–0·48	0·70–0·90	0·040	0·040	0·20–0·35	3·25–3·75	A2345
2515	,, ,,	0·12–0·17	0·40–0·60	0·040	0·040	0·20–0·35	4·75–5·25	A2515

TABLE XXXIX.

S.A.E. NICKEL CHROMIUM STEELS.

S.A.E. No.	Title of Specification.	Chemical Composition.							Corresponding A.I.S.I. No.
		Carbon, per cent.	Manganese, per cent.	Phosphorus, per cent.	Sulphur, per cent.	Silicon, per cent.	Nickel, per cent.	Chromium per cent.	
				Max.	Max.				
3115	Nickel-Chromium Steel	0·13–0·18	0·40–0·60	0·040	0·040	0·20–0·35	1·10–1·40	0·55–0·75	A3115
3120	do.	0·17–0·22	0·60–0·80	0·040	0·040	0·20–0·35	1·10–1·40	0·55–0·75	A3120
3130	do.	0·28–0·33	0·60–0·80	0·040	0·040	0·20–0·35	1·10–1·40	0·55–0·75	A3130
3135	do.	0·33–0·38	0·60–0·80	0·040	0·040	0·20–0·35	1·10–1·40	0·55–0·75	A3135
3140	do.	0·38–0·43	0·70–0·90	0·040	0·040	0·20–0·35	1·10–1·40	0·55–0·75	A3140
X3140 3141	do.	0·38–0·43	0·70–0·90	0·040	0·040	0·20–0·35	1·10–1·40	0·70–0·90	A3141
3145	do.	0·43–0·48	0·70–0·90	0·040	0·040	0·20–0·35	1·10–1·40	0·70–0·90	A3145
3150	do.	0·48–0·53	0·70–0·90	0·040	0·040	0·20–0·35	1·10–1·40	0·70–0·90	A3150
3240	do.	0·38–0·45	0·40–0·60	0·040	0·040	0·20–0·35	1·65–2·0	0·90–1·20	A3240
3312 * 3310	do.	0·08–0·13	0·45–0·60	0·040	0·040	0·20–0·35	3·25–3·75	1·40–1·75	E3310

* *Electric Furnace Steel.*

TABLE XL.

S.A.E. MOLYBDENUM STEELS.

S.A.E. No.	Title of Specification.	Chemical Composition.						Corresponding A.I.S.I. No.
		Carbon, per cent.	Manganese, per cent.	Silicon, per cent.	Nickel, per cent.	Chromium, per cent.	Molybdenum, per cent.	
4023	Molybdenum Steel	0·20–0·25	0·70–0·90	0·20–0·35	—	—	0·20–0·30	A4023
4027	do.	0·25–0·30	0·70–0·90	0·20–0·35	—	—	0·20–0·30	A4027
4032	do.	0·30–0·35	0·70–0·90	0·20–0·35	—	—	0·20–0·30	A4032
4037	do.	0·35–0·40	0·75–1·00	0·20–0·35	—	—	0·20–0·30	A4037
4042	do.	0·40–0·45	0·75–1·00	0·20–0·35	—	—	0·20–0·30	A4042
4047	do.	0·45–0·50	0·75–1·00	0·20–0·35	—	—	0·20–0·30	A4047
4063	do.	0·60–0·67	0·75–1·00	0·20–0·35	—	—	0·20–0·30	A4063
4068	do.	0·64–0·72	0·75–1·00	0·20–0·35	—	—	0·20–0·30	A4068
4119	do.	0·17–0·22	0·70–0·90	0·20–0·35	—	0·40–0·60	0·20–0·30	A4119
4125	do.	0·23–0·28	0·70–0·90	0·20–0·35	—	0·40–0·60	0·20–0·30	A4125
X4130 4130	do.	0·28–0·33	0·40–0·60	0·20–0·35	—	0·80–1·00	0·15–0·25	A4130
4137	do.	0·35–0·40	0·70–0·90	0·20–0·35	—	0·80–1·00	0·15–0·25	A4137
4140	do.	0·38–0·43	0·75–1·00	0·20–0·35	—	0·80–1·00	0·15–0·25	A4140
4145	do.	0·43–0·48	0·75–1·00	0·20–0·35	—	0·80–1·00	0·15–0·25	A4145
4150	do.	0·46–0·53	0·75–1·00	0·20–0·35	—	0·80–1·00	0·15–0·25	A4150
4320	do.	0·17–0·22	0·45–0·65	0·20–0·35	1·65–2·00	0·40–0·60	0·20–0·30	A4320
X4340 4340	do.	0·38–0·43	0·60–0·80	0·20–0·35	1·65–2·00	0·70–0·90	0·20–0·30	A4340
4615	do.	0·13–0·18	0·45–0·65	0·20–0·35	1·65–2·00	—	0·20–0·30	A4615
4620	do.	0·17–0·22	0·45–0·65	0·20–0·35	1·65–2·00	—	0·20–0·30	A4620
4640	do.	0·38–0·43	0·60–0·80	0·20–0·35	1·65–2·00	—	0·20–0·30	A4640
4815	do.	0·13–0·18	0·40–0·60	0·20–0·35	3·25–3·75	—	0·20–0·30	A4815
4820	do.	0·18–0·23	0·50–0·70	0·20–0·35	3·25–3·75	—	0·20–0·30	A4820

Phosphorus and Sulphur, each 0·04 per cent. max. throughout above series.

TABLE XLI.

S.A.E. CHROMIUM STEELS.

S.A.E. No.	Title of Specification.	Chemical Composition.						Corresponding A.I.S.I. No.
		Carbon, per cent.	Manganese, per cent.	Phosphorus, per cent.	Sulphur, per cent.	Silicon, per cent.	Chromium, per cent.	
				Max.	Max.			
5120	Chromium Steel	0·17–0·22	0·70–0·90	0·040	0·040	0·20–0·35	0·70–0·90	A5120
5140	,, ,,	0·38–0·43	0·70–0·90	0·040	0·040	0·20–0·35	0·70–0·90	A5140
5150	,, ,,	0·48–0·55	0·70–0·90	0·040	0·040	0·20–0·35	0·70–0·90	A5150
52100*	,, ,,	0·95–1·10	0·30–0·50	0·040	0·040	0·20–0·35	1·20–1·50	E52100

* *Electric Furnace Steel.*

TABLE XLII.

S.A.E. CHROMIUM-VANADIUM STEEL.

S.A.E. No.	Title of Specification.	Chemical Composition.						Vanadium, per cent.
		Carbon, per cent.	Manganese, per cent.	Phosphorus, per cent.	Sulphur, per cent.	Silicon, per cent.	Chromium, per cent.	
				Max.	Max.			Min.
6150	Chromium-Vanadium Steel	0·48–0·55	0·65–0·90	0·040	0·040	0·20–0·35	0·80–1·10	0·15

TABLE

S.A.E. SILICON-MANGANESE STEEL.

S.A.E. No.	Title of Specification.	Chemical Composition.					Corresponding A.I.S.I. No.
		Carbon per cent.	Manganese, per cent.	Phosphorus, per cent.	Sulphur, per cent.	Silicon, per cent.	
				Max.	Max.		
9260	Silicon-Manganese Steel	0·55–0·65	0·70–0·90	0·040	0·040	1·80–2·20	A9260

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TABLE XLIV.

BRITISH STANDARD SPECIFICATIONS FOR STEEL BARS AND FORGINGS. D.T.D. SPECIFICATIONS.

D.T.D. No.	Title of Specification	Carbon %	Silicon %	Manganese %	Sulphur %	Phosphorus %	Nickel %	Chromium %	Vanadium %	Molybdenum %	Tungsten %	Ultimate Tensile Stress Tons/sq.in.	Minimum Proof Stress Tons/sq.in.	Elongation %	Brinell Hardness	Izod ft-lb.	Nickel rattle test see foot-note
		Max.	Max.	Max.	Max.	Max.	Max.	Max.	Max.	Max.	Max.	Min.	Min.	Min.		Min.	
6A	Cobalt-Chromium Valve Steel	1·10–1·60	0·75	0·20–0·60		0·03	0·50	11·0–14·0 7·5–8·5	†	0·5–1·0	2·0–4·0	32 Min.			228–277 235–286		Bend test, 45° round 0·10 in. rad.
13B	Silicon-Chromium Valve Steel Forgings or Stampings	0·40–0·50	3·25–3·75	0·40–0·60			0·50	7·5–8·5									×
40B	High Nickel High Chromium Steel Valve Forgings	0·85–0·50	1·0–2·50	1·50			10·00 Min.	12·0–16·0				200 Max.				15	×	
53	Low Tensile Steel Bar (non-corrodible)	0·10 Max.	0·50		0·05		1·00	12·0 Min.		0·10–0·25		25–34		20	131–170	33	×
87	Steel for Nitrogen Hardening‡	0·35–0·45	0·45	0·65	0·05	0·04	0·25	1·4–1·8	*		*	55–65		17	241–302	33	×
115	Silico-Manganese Steel Bars	0·45–0·56	1·60–2·10	0·80–1·80	0·05	0·05	0·40	0·10				90–105		10	401–444		×
126A	"40" Carbon Steel (for welding)	0·30 Max.	0·30	1·75	0·05	0·05	1·00					40–55	31	20	157–248	33	×
158	Bright Steel Bars for Pins and High Tensile Bolts	0·35–0·60	0·30	0·50–0·90	0·05	0·05	1·20*					55–65		20	241–293	33	×
176A	Chromium Nickel Steel§ (non-corrodible)	0·20	0·20 Min.	1·00	0·05	0·05	8·0–20·0	12·0 Min.	*		*	35 Min.	15	30		30	×
188A	55–65-ton Carbon Steel	0·25–0·40	0·35	1·30–1·80	0·05	0·05				0·2–0·55		55–65		18	241–311	40	×
225	55–65-ton Nickel Chromium Molybdenum Steel (Nitrogen hardening)	0·25–0·85	0·35	1·00	0·05	0·05	1·00	0·5–1·5	0·25*	0·9–1·5	1·0*	55–65		18	248–293	35	× ×
247	High Thermal Expansion Steel (valve seats)	0·70 Max.	0·50	3·50–5·50	0·05	0·05	11·0–14·0	8·0 Min.	0·25*	0·50*	1·0*	40 Min.		25	672		—
261	Alloy Steel Bars	0·85–0·95	0·50	1·70	0·05	0·05	0·25*	0·60* 17·0 Min.	0·85*	0·60*	1·0* 2·0–4·0				902 Max.	30	×
282	High Chromium Steel Valve Forgings	0·80–0·45	1·0–2·50	1·00	0·05	0·05	6·0–10·0	1·4–2·50	0·25*	0·15–0·5	1·0*	55–65	43	18	248–302	35	×
286A	55–70-ton Chromium Molybdenum Steel (Nitrogen hardening)	0·82–0·48	0·50	0·65	0·05	0·05	0·40										×
299	Mild Steel Bars, Forgings and Tubes	0·15 Min.	0·30	0·60	0·05	0·05	0·30		0·25*	0·3–0·7	1·0*	60–70	45§§	17	120 Max. 269–321	35	×
306	60–70 Chromium Molybdenum Steel (Nitrogen hardening)	0·15–0·35	0·35	0·65	0·05	0·05	0·30	2·5–3·5								12	×
311	Silicon Chromium Valve Steel Forgings or Stampings	0·55–0·65	1·40–1·70	0·30–0·80	0·05	0·05	0·50	5·75–6·75	0·25*	0·3–0·7	1·0*	45–65	30	20	195–277 207–255	43	×
317A	45–55-ton Chromium Molybdenum Steel§ (Nitrogen hardening)	0·15–0·35	0·30	0·65	0·05	0·05	0·30	2·5–3·5	0·25*	0·8–0·7	1·0*						×
381	80–90-ton Nickel Chromium Steel	0·25–0·40	0·30	0·70	0·05	0·05	8·0–4·50	0·75–1·5	0·25*	0·65*		80–90	70	14	388–415	25	× × × × × ×
398	Low Tensile Nickel Chromium Steel Bolts	0·25–0·35	Composi-tion to be	0·60–0·90	0·05 agreed.	0·05	0·80	0·8–1·2				35 Min. 55–65		18		40	
401	High Tensile Cold Headed Steel Bolts		Chemical	0·60–1·20			2·0	12·0 Min.				55–65	43§§	18	241–311	40	
461	55–85-ton 1% Chromium Steel	0·85–0·45	0·35	1·00	0·05	0·05		0·9–1·5				55–70	48§§	15	241–311	40	
463	55–70-ton Non-corrodible Steel	0·15–0·85	1·00	0·65								55–65	48§§	18	241–311	25 Min.	
470	55–65-ton Chromium Molybdenum Steel	0·85–0·45	0·35	0·5–0·8	0·05	0·05				0·25–0·4						40	
473	85-ton Nickel Chromium Molybdenum Steel																
	For sizes up to 1¼ in.	0·85–0·45	0·10–0·35	0·5–0·7	0·05	0·05	1·8–2·80	0·9–1·4		0·2–0·4*		40–50	30	14 (for sizes <1¼ in.) 12 (for sizes >1¼ in.)	332–402	25	×
	For sizes over 1¼ in. but not exceeding 2½ in.	0·27–0·35	0·10–0·86	0·5–0·7	0·05	0·05	2·8–3·8	0·5–0·8		0·4–0·7		75–85	65§§				
480	For sizes over 2½ in. 55–65-ton 1½% Nickel Chromium Molyb-denum Steel	0·86–0·44	0·10–0·85	0·5–0·7	0·05	0·05	2·8–3·8 1·0–2·0	0·75–1·5 0·5–1·0		0·4–0·7 0·2–0·4		55–85	48§§	18	241–311	40	×
490	55–65-ton 2½% Nickel Chromium Molyb-denum Steel (Medium Carbon)	0·25–0·85	0·35	0·4–0·8	0·05	0·05	2·0–3·0	0·5–1·0		0·4–0·7		55–65	48§§	18	241–311	40	× ×
500	80–90-ton 2½% Nickel Chromium Molyb-denum Steel (High Carbon)	0·85–0·45	0·35	0·4–0·8	0·05	0·05	2·0–3·0	0·50–1·0		0·4–0·7		80–90	70§§	14 (for sizes <4 in.) 12 (for sizes ≥4 in.)	352–429	25	×
510	40–50-ton Manganese Nickel Molybdenum Steel	0·25 Max.	0·35	1·4–1·8	0·05	0·05	0·3–0·7			0·2–0·4		40–50	28§§	22	163–241	40	×
519	3% Nickel Chromium Case-hardening Steel	0·18 Max.	0·35	0·80–0·60	0·05	0·05	8·0–8·75	0·80–1·10				65 Min.	28	18		30	×
525	45–65-ton Non-corrodible Steel (Free Machining)	0·85 Max.	1·00	1·00	0·75	0·05	1·0	12·0 Min.				45–55		15		25.	

(Selenium 0·6 max., Zirconium 0·6 max., Molybdenum 0·6 max., Lead 0·35 max. Total not more than 1·0.)

* Optional.
† Cobalt 3·0 to 5·0 per cent.

‡ Aluminium 0·90–1·30 per cent.
|| Copper and Titanium optional.

§ Aluminium (if present) 0·60 per cent. max.
§§ For information only.

N.A. cross indicates that a nicked fracture test is specified.

TABLE XLV.

### BRITISH STANDARD SPECIFICATIONS FOR STEEL, BARS AND FORGINGS. BRITISH AIRCRAFT STANDARDS

B.S. No.	Title of Specification	Carbon %	Silicon % Max	Manganese % Max	Sulphur % Max	Phosphorus % Max	Nickel % Max	Chromium % Max	Vanadium % Max	Molybdenum % Max	Tungsten % Max	Ultimate Tensile Stress Tons/sq.in.	0·1% Proof Stress Tons/sq.in. Min	Elongation % Min	Reduction of Area % Min	Izod ft.-lb. Min	Brinell Hardness	Nicked fracture test (see footnote)
351	Bright Steel Bars	0·15-0·40	0·30	0·50-0·90	0·05	0·05	—	—	—	—	—	35-45	—	15·0	40	40 (up to and including ¾ in.); 25 (Up to and including 1¼ in.); 20 (over 1½ in.)	—	—
	35-ton Alloy Steel Bars																	
382	"40" Carbon Steel (normalised)	0·35-0·45	0·30	1·20	0·05	0·05	—	—	—	—	—	35-45	20	18·0	50	40	241-293	X
386	55-65-ton Nickel Chromium Steel	0·25-0·35	0·30	0·45-0·70	0·05	0·05	1·00*	0·5-1·0	0·25*	0·65*	1·00*	55-65	43	20·0	—	20	140-201	X
4S11	Carbon Case-hardening Steel	0·10-0·18	0·30	0·90	0·05	0·05	—	—	—	—	—	45-60	—	18·0	50	40	248-302	X
S15	3% Nickel Case-hardening Steel	0·10-0·15	0·30	0·20-0·60	0·05	0·05	2·75-3·75	0·30	—	—	—	55-65	—	20·0	45	40	—	X
S21	"20" Carbon Steel (suitable for welding)	0·25 Max.	0·30	1·00	0·05	0·05	—	—	—	—	—	32 Min.	—	18·0	—	—	—	—
3S24	Bright Steel Bars for keys	0·50-0·70	0·30	0·50-0·90	0·05	0·05	0·30*	—	—	—	—	25-35	—	12·5	25	15	229-293	X
2S28	Air Hardening Nickel Chromium Steel	0·25-0·32	0·30	0·35-0·60	0·05	—	3·75-4·50	1·0-1·5	0·25*	0·65*	1·00*	100 Min.	—	12·0	50	45	444 Min.	—
S61	35-45-ton High Chromium Steel (non-corroding)	0·15 Max.	0·30	—	—	—	1·00	12·0 Min.	—	—	—	35-45	—	25·0	—	25 (Up to and including 2 in.); 20 (over 2 in.)	152-207	X
4S2	44-52-ton High Chromium Steel (non-corroding)	0·15-0·35	0·50	0·35-0·65	—	—	1·00	12·0 Min.	—	—	—	44-52	—	20·0	45	35 (Up to and including 2 in.); 20 (over 2 in.)	207-235	X
S65	65-ton Nickel-Chromium Steel	0·22-0·28	0·30	0·35-0·65	0·05	0·05	2·75-3·50	1·0-1·4	0·25*	0·65*	1·00*	65-70	25	17·0	40	35	293-321	X
S67	5% Nickel Case-hardening Steel	0·08-0·14	0·30	0·35	0·05	0·05	4·60-5·20	0·10	—	—	—	40-60	—	20·0	45	50	—	X
S68	10% Tungsten Steel	0·55-0·70	—	0·40	—	—	—	3·5 Min.	1·00*	—	14·0 Min.	—	—	—	—	—	277 Max.	—
S69	3½% Nickel Steel	0·35-0·45	0·30	0·50-0·80	0·05	0·05	3·25-3·75	0·30	—	—	—	55-65	—	18·0	50	35	241-293	—
S70	"55" Carbon Steel (normalised)	0·50-0·60	0·30	0·40-0·75	0·05	0·05	—	—	—	—	—	45 Min.	—	18·0	50	—	197-241	X
S71	"30" Carbon Steel (normalised)	0·25-0·35	0·30	1·20	0·05	0·05	—	—	—	—	—	25-35	—	25·0	—	20	109-163	X
2S76	"40" Carbon Steel (hardened and tempered)	0·35-0·45	0·30	1·20	0·05	0·05	1·00*	—	—	—	—	30-40	25	22·0	50	35	174-223	X
S77	"40" Carbon Steel (hardened and tempered)	0·35-0·45	0·30	1·20	0·05	0·05	—	—	—	—	—	30-40	—	25·0	35	25	131-174	X
S79	"55" Carbon Steel (hardened and tempered)	0·50-0·60	0·30	0·40-0·75	0·05	—	—	—	—	—	—	55 Min.	—	15·0	—	25	235-285	—
S80	55-ton High Chromium Steel (non-corroding)	0·25 Max.	0·50	0·70	—	0·05	1·00	16·0-20·0	—	—	—	55 Min.	—	15·0	—	35	241 Min.	X
2S81	65-75-ton Nickel Chromium Steel	0·29-0·35	0·30	0·70	0·05	0·05	3·0-3·75	0·5-1·30	0·25*	0·50*	1·00*	65-75	55	16·0	35	35	255-285	X
S82	Nickel Chromium Case-hardening Steel	0·18 Max.	0·30	0·50	0·05	0·05	4·0-4·50	1·0-1·60	0·25*	0·50*	1·00*	85 Min.	—	12·0	—	30	—	X
S90	High tensile 5% Nickel Case-hardening Steel	0·16 Max.	0·30	0·60	0·05	0·05	4·5-5·50	0·30	0·25*	0·50*	1·00*	65 Min.	—	13·0	—	—	293-341	X

* Optional.   X A cross indicates that a nicked fracture test is specified.

(v)

(vii)

How do you Maintain
Consistent Temperature of
Quench Baths?